HIGH YIELD STRENGTH CAST STEEL WITH IMPROVED WELDABILITY

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NSRP NO. 0326
MAY 1991

HIGH YIELD STRENGTH CAST STEEL WITH IMPROVED WELDABILITY

FINAL REPORT

Submitted to:

The SP-7 Panel
of
The Ship Production Committee
of
The National Shipbuilding Research Program

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March 18, 1991

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The National Shipbuilding Research Program (NSRP) has been engaged in research related to improvements in shipbuilding in the U.S. since 1973. The program is a cooperative effort involving conrnercial and U.S. Naval shipyards and related agencies, industries and educational institutions.

Since the inception of the NATIONAL SHIPBUILDING RESEARCH PROGRAM in 1973, R&D projects have made significant contributions to shipbuilding in the areas of facilities, environmental effects, outfitting, production aids, design and production integration, welding, industrial engineering, education and training, flexible automation and coatings. A bibliography and library of NSRP reports is maintained at the University of Michigan, Transportation Research Institute, Ann Arbor, Michigsan.

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FOREWORD

Significant cost and weight savings in shipbuilding have resulted from introduction of more weldable steels, such as HSLA 80 and HSLA 100 plates and beams, in the decade of the '80's. This has given notice to the fact that improvements in shipbuilding productivity and cost savings may be found in better materials as well as in faster welding processes and automation.

The SP 7 panel of the National Shipbuilding Research Program recognized several years ago that a corollary improvement in weldability Of high yield strength steel castings would provide technical or economic advantages not available from current state of the art steel castings. For example, if the toughness and yield strength of HY 80 could be retained in a steel casting alloy which forgives the need to preheat prior to installation or repan welding, then both technical and economic advantages would accrue.

In this project to develop a more weldable replacement for HY 80, ESCO Corporation, Portland, Oregon teamed with Oregon Graduate Center, Beaverton, Oregon to cast and to test the alloy variations which their combined knowledge and experience indicated are likely paths to the objectives.

Robin Churchill and his associates at ESCO Corporation developed the initial approach to the task which included as a first premise the possibility of improving weldability in a NiCrMo system by reducing carbon content. They drew from many years of experience with HY 80 and many other alloy systems and were able to go, with a minimal number of iterations of chemistry, to four "optimized" melts which met the initial objectives of the project. Churchill and his associates at ESCO produced the experimental melts. Dr.Jack Devletian and his associates at Oregon Graduate center provided metallurgical, and mechanical testing support throughout the project as well as subcontract administrative support. Most of the welding and mechanical testing was performed at OGC. The project report was the joint effort of Robin Churchill, Jack Devletian and Daya Singh.

The initial group of five trial melts were variations of HY 80, HY 130 and HSLA 80 chemistry with emphasis on low carbon and careful control of nickel and other constituents. Test results were used to guide selection of chemistry and thermal treatment for another group of four optimized combinations of chemistry and heat treatment. Castings with chemistry similar to HSLA 80 did not exhibit sufficient low temperature toughness to warrant further consideration as a candidate casting alloy.

The four optimized melts produced two which met all of the mechanical property requirements of Mil-S-23008 for HY 80 castings as well as the weldability tests. Apparently the reduction of carbon to about 0.045% and some increase of nickel allowed good hardness levels in heat affected zones while preventing HAZ embrittlement. Using Mil 100-S1 wire, the weld metal in all four optimized castings passed the Gas Metal Arc (GMA) welded Controlled Thermal Severity (CTS) test for underbead cracking. Mil 100-S1 is the wire in general use in U.S shipyards for HY 80 GMA welding. In the more severe Tekken retraint test, two samples welded with no preheat using the flux core process showed cracks which were limited to the weld metal. The flux core filler metal used was TM-101. Two alloys showed no adverse effects of elimination of preheat in any of the restraint tests. On the other hand, the HY 80 reference casting exhibited typical underbead cracking when welded with no preheat, regardless of the weld process used.

The new casting alloys produced in this project appear not only to be capable of meeting mechanical requirements of HY 80 but also to be weldable without preheat. The incremental increase in alloy cost resulting from the enriched chemistry (over 5% nickel) and a possible need for double tempering is expected to be offset by savings resulting from lower repair costs in production and the reduction or elimination of preheat costs in shipbuilding.

This project was financially limited to producing and screening the experimental casting alloys. This limited scope could not provide for the much more extensive weldability (and other) testing required to qualify a new alloy for use in ship production however, the positive outcome of this sequence of alloy development and testing justifies taking another step toward the objective.

The SP 7 panel considers it appropriate to request support for follow up work in which larger heats will be melted to evaluate producibility on a larger scale and to provide for more extensive testing of foundry producibility, thermal effects, mechanical property tests, re-heat cracking susceptibility and no-preheat weldability. It is expected that some of the new SMAW and hare wire filler metalfi now under development for HSLA 80, HSLA 100 and HY 130 in other SP 7 and U.S. Navy projects will be available for the no-preheat weldability testing of the castings to be produced in the follow up SP 7 project. Results of these projects may combine to produce interactive and complementary results. Explosion bulge tests will probably be performed on cast plates.

A significant contribution to the development of a new, more weldable high yield strength casting for marine applications has been achieved by this project at a modest cost. That achievement is primarily attributable to the professionalism and efficiency of the principal investigators and the availability of the facilities of ESCO Corporation and Oregon Graduate Center.

Finally, as will be seen in the project report, development of an 80 KSI yield strength steel casting alloy with major improvements in weldability, is achievable without a major cost increase in alloy chemistry. It is hoped that the follow up work will show that cost savings already realized in shipbuilding in welding HSLA 80 and HSLA 100 plates can be extended to include welding of shaft struts, rudders, sonar dome structures and other steel castings without preheat.

O.J. Davis.Program Manager
SP 7 Panel
May, 1990

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Note: This report was performed under the National Shipbuilding Research Program. It was initiated by the SP 7 panel of the Ship Production Committee of the Society of Naval Architects and Marine Engineers. It was funded with joint suPport of the U.S.Navy and the U.S. Maritime Administration under MARAD contract No. DTMA 91-84-41028 with Ingalls Shipbuilding Inc. The project was performed under subcontract* to ESCO Corporation and Oregon Graduate Center under the SP 7 Chairmanship of Lee Kvidahl and Program Management of O.J. Davis.

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OBJECTIVE

The objective of this investigation was to evaluate a class of very low carbon quenched-&-tempered cast steels as possible replacements for cast HY-80 per MIL-S-23008. To be acceptable. the alloy(s) must:

1. meet the mechanical property requirements of cast HY-80, and

2. be weldable without preheating.

ABSTRACT

n order to complement wrought HSLA-80, it was decided to attempt to develop a substitute for cast EY-80 which would permit welding without the need for preheating. A number of heats of very low-carbon, higher-nickel modifications of HY-80 and low-carbon modifications Of HY-130 were produced and cast into test blocks. Heat treatment studies were performed and mechanical properties were evaluated. In addition, weldability tests were performed to assess susceptibility to hydrogen-assisted cracking when welding without preheating. The experimental alloys appear to be capable of meeting the mechanical property requirements of HY-80 in section thicknesses up to at least 12 in. The weldability tests revealed no HAZ cracking of the experimental alloys when welded without preheat while cast HY-8fl was found to suffer severe underbead cracking under the same conditions.

INTRODUCTION

In recent years, there has been an increased need in the shipbuilding industry to decrease weight by making greater use of higher strength steels. High strength structural steels, of course, have been available for many years. One such alloy, HY-80, has a long history of use in naval ship construction. Unfortunately, HY-80 must generally be preheated for welding and this significantly increases welding costs. The desire to reduce welding costs has led to the development of "HSLA-80" which is a wrought precipitation hardening steel based on ASTM A71O. This material has mechanical properties comparable to those of HY-80, but does not require the costly practice of preheating for welding. HSLA-80 is presently being utilized extensively as a replacement for wrought HY-80 on U.S. Navy surface ships.

Traditionally, structural steels with yield strength levels greater than about 65 KSI have been primarily martensitic or martensitic/bainitic materials. Materials with these types of microstructure are capable of attaining excellent combinations of strength and low-temperature toughness. Unfortunately, the compositions which are ultimately responsible for the good characteristics of these materials also cause susceptibility to hydrogen-assisted cracking in the HAZ’S of welds. To avoid cracking, it is necessary to preheat these materials so as to allow the HAZ to transform to softer, less crack-sensitive microstructure.

To achieve its superior weldability, HSLA-80 does not rely on the conventional quenched-and-tempered technology employed in HY-80. HSLA-80 has an essentially ferritic microstructure and achieves its strength through precipitation hardening rather than a martensitic transformation (Refs. 1, 2 & 3) . This allows the material to be produced with a much lower carbon content (0.07 max.) . This reduces the maximum attainable hardness in the HAZ's of welds, which in turn reduces the tendency for hydrogen-assisted cracking. By the use of microalloying (and, in some cases, thermo-mechanical processing) to produce fine grain size, good toughness can also be obtained in HSLA-80, in spite of its non-martensitic microstructure.

While HSLA-80 has largely replaced HY-80 in wrought forms, no substitute for cast HY-80 has yet been developed. Thus , preheating (250°F to 350°F) is still required for welding castings. In 1985, Mr. O. J. Davis of Ingalls Shipbuilding contacted ESCO Corporation in order to discuss potential substitutes for cast HY-80. Mr. Davis described the successes which had been achieved with replacing HY-80 with HSLA-80 (known as modified ASTM A71O at that time) and asked if ESCO would consider evaluating the material in the cast form. ESCO agreed to do so and performed a small investigation. A section of an as-cast A710 ingot was obtained from a wrought steel producer for evaluation. In addition, a small air-melt induction heat of the material was produced by ESCO. Heat treatment studies were performed and mechanical properties were evaluated. It was found that while appropriate strength levels could be attained, the material exhibited poor low-temperature toughness (Ref. 4).

A variety of cast low-carbon HSLA steels have been introduced in recent years (Ref.5). Many of these are readily welded without the need for preheating. As with wrought HSLA-80, several of these alloys employ precipitation hardening to achieve the desired strength levels. (Of course, since they are cast materials, they are not subjected to the thermo-mechanical treatments used to produce some forms of wrought HSLA-80.) While several of the new cast alloys can meet a strength level of 80 ksi, none of them possesses adequate low-temperature impact toughness to serve as a substitute for cast HY-80. In addition to the toughness deficiencies, evidence is growing that the heavy section weldability of these types of steels may be suspect.

Within the last few years, a new embrittlement phenomenon known as "localized brittle zones" (LBZ’s) has been plaguing HSLA steels developed for offshore oil platform construction (Ref. 6). This problem appears to be a special case of the embrittlement associated with "reheat cracking." The most serious forms of this type of embrittlement occur in the coarse-grained HAz's of multipass welds in steels containing microalloying additions such as columbium and vanadium. It also is expected to occur in steels which are precipitation hardened with copper.

The preceding information has important implications regarding any attempt at developing a more weldable replacement for cast HY-80. The poor low-temperature toughness of cast HSLA-80 and the similar commercial cast steels suggests that no alloy of this type would be capable of meeting the mechanical property requirements of cast HY-80. In order to achieve a suitable combination of strength and toughness, a martensitic or martensiticllower-bainitic microstructure (much like that of HY-80 itself) will probably be required. Any improvement in weldability, without the requirement for preheating, then, will have to be achieved through reducing the carbon content below that of HY-80. This will reduce the maximum possible HAZ hardness and, accordingly, should reduce the tendency for hydrogen-assisted cracking. To achieve the desired microstructure and tempering resistance with the lowered carbon content, it will probably be necessary to increase the nickel content beyond the nominal 3% level of HY-80. Since the cited literature suggests that the use of precipitation hardening additions may lead to HAZ embrittlement and reheat cracking in heavy section nultipass welds, this strengthening mechanism should be avoided.

The weldability of hardenable steels is commonly expressed in terms of “carbon equivalent” (C.E.) numbers. Carbon equivalent expressions take into account the tendency to form hard trans-formation products in weld HAZ’S (hardenability) as well as the expected hardness of those transformation products (carbon content) . Hiaher carbon equivalent numbers aenerallv indicate greater susceptibility to hydrogen assisted cricking in the HAZ’S of welds. However, it is recognized that at any particular C.E. value, higher carbon levels produce greater susceptibility to cracking (Ref. 7). This is illustrated in Figure 1. As can be seen, low C.E. levels provide good weldability for the range of carbon contents encountered in most common structural steel grades. Higher C.E. levels tend to promote cracking. It should be noted, however, that at low carbon contents, (e.g. <0.1%), cracking resistance is good even at relatively high C.E. levels. If it were possible to develop an alloy which could attain the desired mechanical properties at very low carbon levels, (e.g. 0.05%), the material would be expected to exhibit good resistance to HAZ hydrogen cracking.

HY-80 is currently used to produce castings of a wide range of sizes. A number of typical HY-80 castings are shown in Fig. 2. Casting weights may range from a few pounds to 25 tons or more. Section thicknesses range from a fraction of .an inch through at least 17 inches. Any replacement for cast HY-80 would need to be capable of attaining the desired mechanical properties in very heavy sections. In quenched-and-tempered steels, this capability is insured by alloying sufficiently to allow formation of substantially-martensitic or at least martensiticflower bainitic microstructures in the desired section sizes. When appropriately tempered, these microstructure are capable of giving good combinations of strength and toughness. In general, the hardenability must be sufficient to suppress formation of proeutectoid ferrite and the coarser ferrite/carbide aggregates (pearlite and upper bainite) since these microstructure lead to poor toughness.

In designing a low-carbon replacement for HY-80, it would be necessary to alloy the material sufficiently to give adequate hardenability to avoid formation of ferrite, pearlite and upper bainite in heavy sections. Combinations of chromium molybdenum and nickel are very effective for this purpose. While chromium and molybdenum are much more efficient hardenability-promoters than nickel, the triple combination provides a synergistic effect. In addition, when added in relatively large percentages, nickel tends to greatly retard formation of ferrite, pearlite and upper bainite. It also tends to lower the bainite-start temperature which causes any bainite which does form to have a fine carbide distribution. These effects are illustrated in the CCT diagrams of two low carbon Cr-Ni-Mo steels shown in Figure 3 (Ref. 8). All of these effects of nickel additions are of great benefit in attaining good toughness.

Nickel additions would also be important in permitting a fully-austenitic structure to be achieved during heat treating. At very low carbon levels, the gamma-loop tends to become very restricted, making it difficult to completely austenitize the material. If ferrite pools are present at the austenitizing temperature, they will be present after quenching as well and this will drastically impair low temperature toughness. Since nickel is an austenite stabilizer, additions of this element can compensate for low carbon levels.

The concept of improving the weldability of HY-80 by reducing the carbon content to low levels and increasing the nickel content has a very close parallel in cast martensitic stainless steels. The cast 12% Cr alloy CA-15 was notoriously difficult to weld. It was found that by reducing the carbon content below 0.06% and adding approximately 4% nickel and about 0.5% molybdenum, weldability improved markedly. The new alloy, CA-6NM, has largely replaced CA-15 in a wide variety of applications, most notably those involving heavy sections (hydro-turbine parts, oil industry valves etc.).

If a substitute for cast HY-80 were to be developed, the material would need to be capable of meeting the same mechanical property requirements specified for HY-80. For military applications, one of the more commonly-used specifications for cast HY-80 is MIL-STD-23008. The specification gives required mechanical property values, gives the sizes of test blocks required to represent castings of a given thickness and indicates the locations where test specimens must be removed from the test block.

Currently, Revisions B and C of MIL-STD-23008 are most frequently specified. During the course of the present study, a new revision, Rev. D was issued. All three revisions specify that test block sizes must be T x T x 6/T, where T is the maximum section thickness of the casting being represented. The required mechanical properties are indicated in Table 1 below:

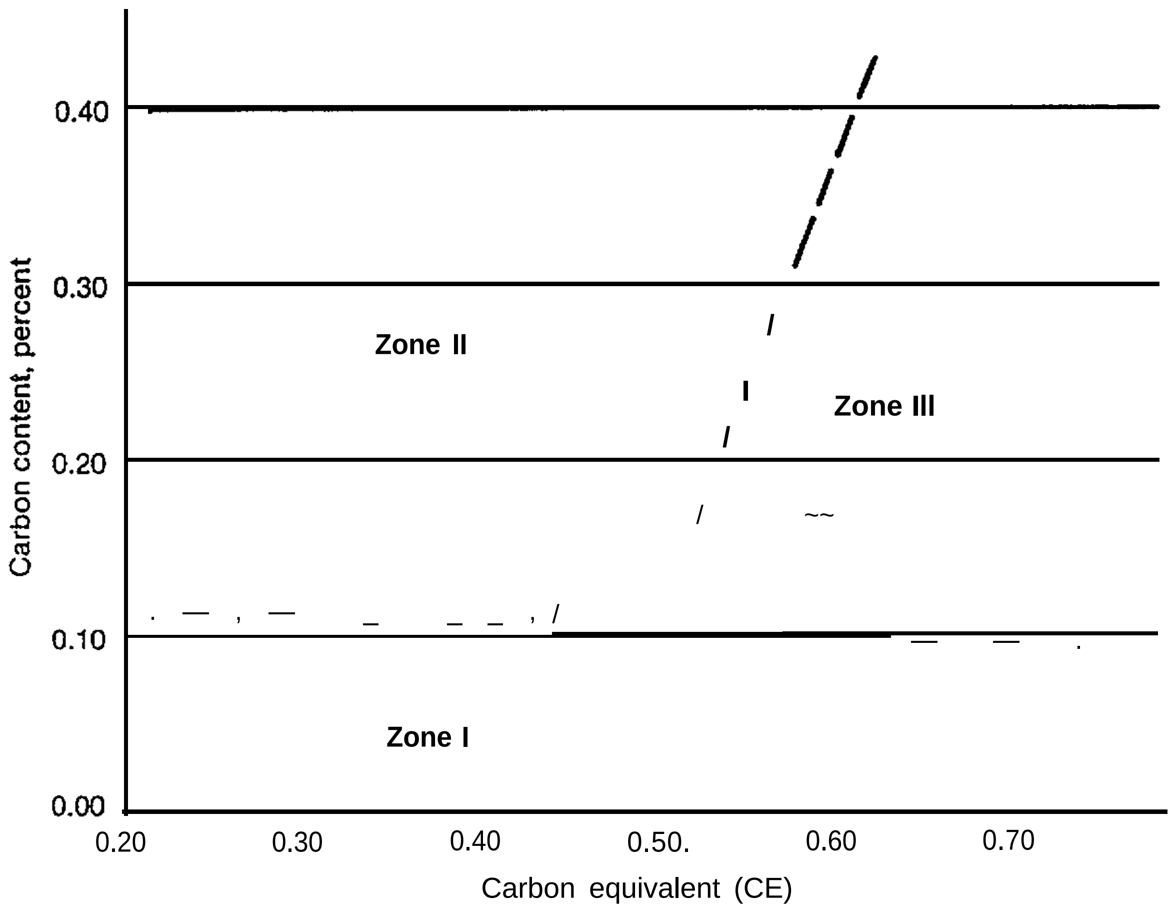
TABLE 1

MECHANICAL PROPERTY REQUIREMENTS FOR CAST HY-80
PER MIL-STD-23008

		REVISION B	REVISION C	REVISION D
YIELD STRENGTH (KSI)		80.0 TO 95.0	80.0 TO 99.5	80.0 TO 99.5
PERCENT ELONGATION		20	20	20
PERCENT REDUCTION OF AREA		35	35	35
Cv (Ft Lbs)	0° F		70	70
	-100 °F	30	50	50

Revisions B and C require that the test specimens be removed a minimum of 1 in. from and surface of the test block. Revision D requires that the specimens be removed at least 1/4 T from the surface. This greater test specimen depth makes Rev. D much more difficult to meet. Compared to testing at l'! from the edge of the tasking, the disadvantages of testinq material from the 1/4 thickness are: (1) the slower cooling rite producing large; dendrite arm spacing and greater microsegregation, and (2) slower cooling rate during the quenching operation (of the quenching and tempering heat treatment) requiring greater hardenability to achieve 80 ksi minimum yield strength while maintaining the required CVN impact toughness. In this research, both locations (specified by revisions B, C and D) were tested for each alloy.

Using the alloy design concepts described above, it was decided to produce and evaluate a number of experimental alloys as potential replacements for cast HY-80.



Note: CE= C+(Mn+Si) /6+(Cr+ Mo+V) /5+ (Ni+Cu)/15
Note 2: See 05.2 (1), (2), or (3), for applicable zone characteristics.

Figure 1.

The effects of carbon content and carbon equivalent number on hydrogen cracking susceptibility.
(From AWS D1.1 Structural Welding Code, Ref. 7).

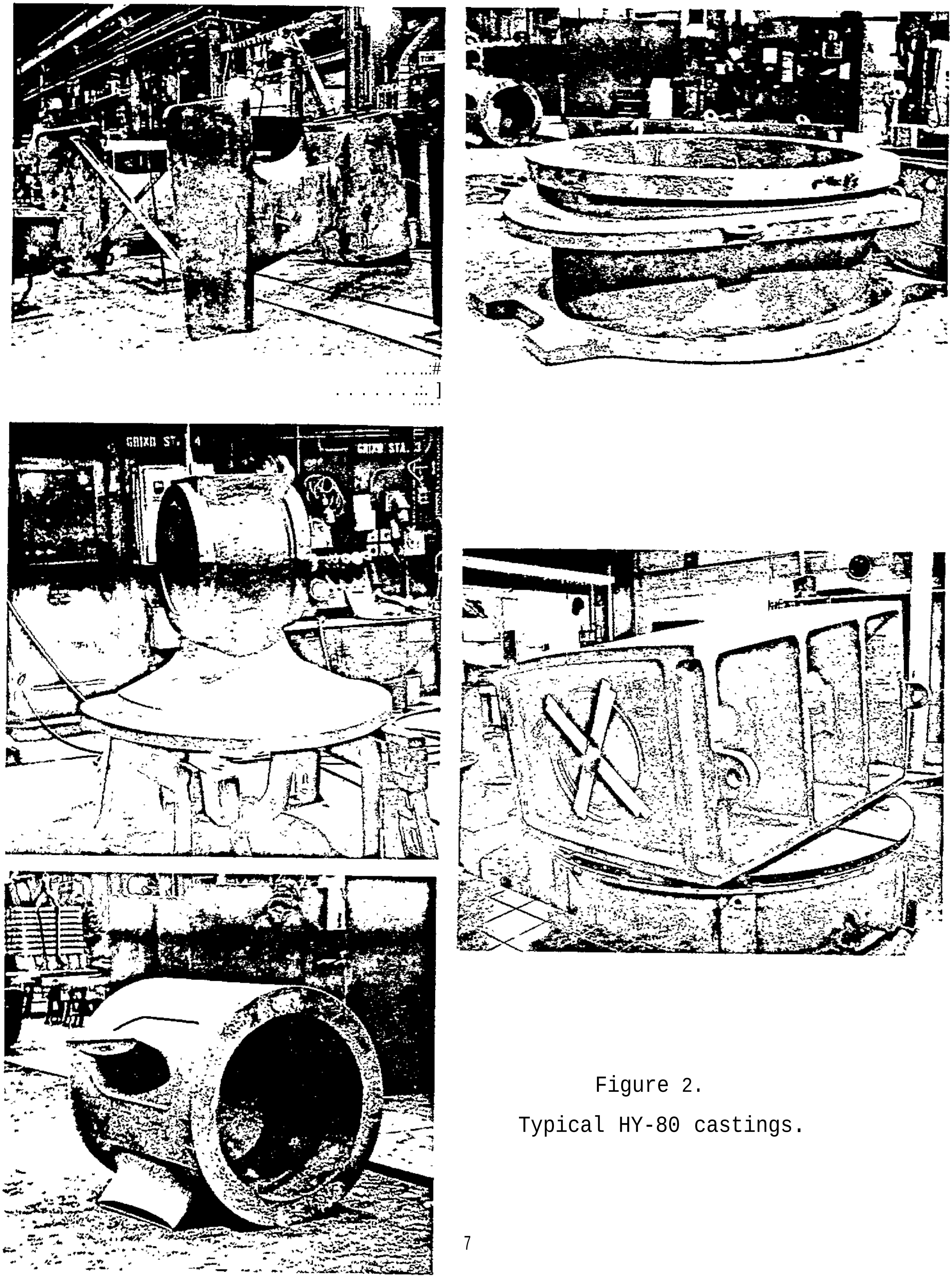


Figure 2.

Typical HY-80 castings.

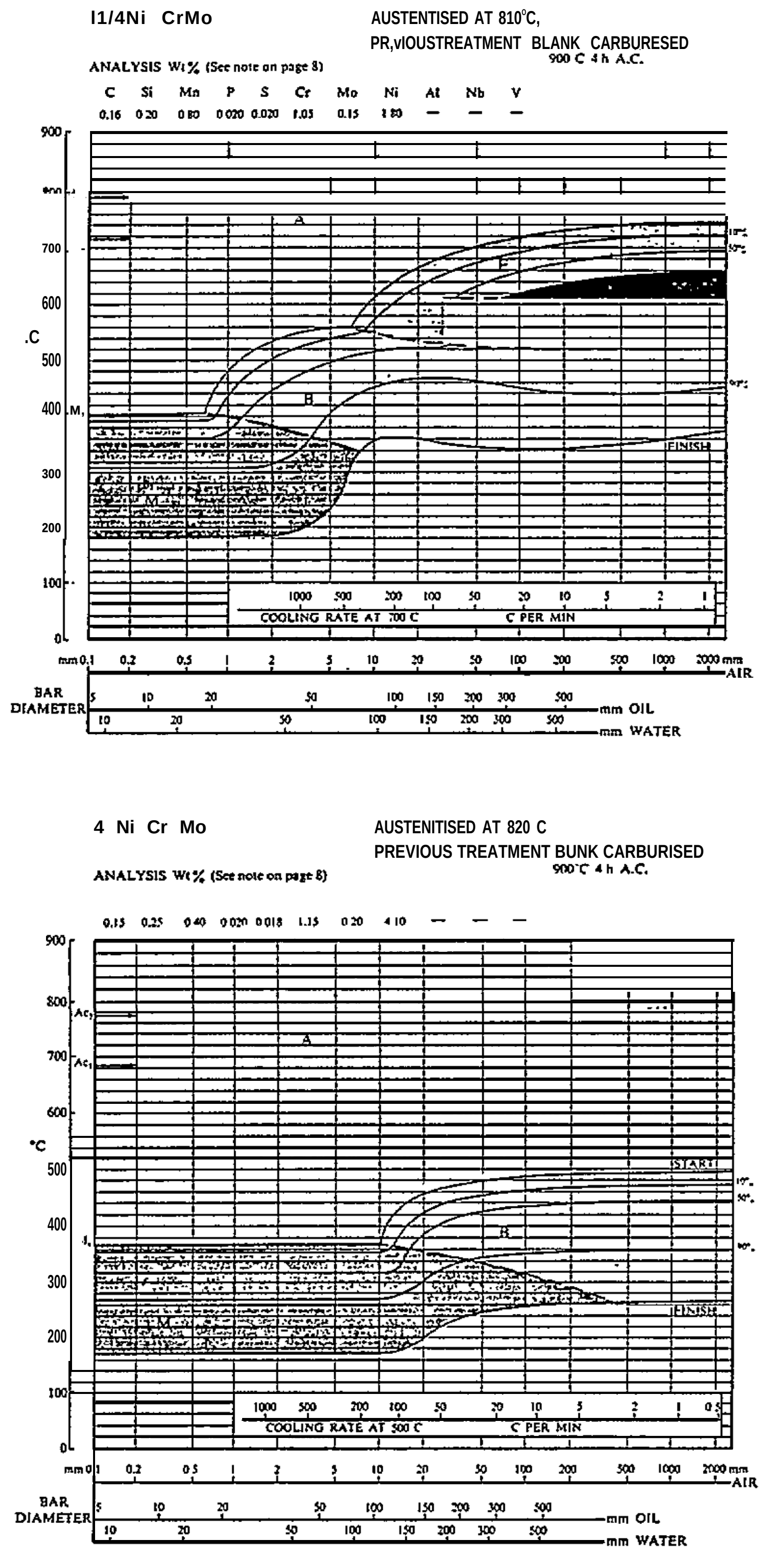


Figure 3.

Continuous cooling diagrams showing the effects of relatively large nickel additions to low carbon low alloy steels.

EXPERIMENTAL PROCEDURES

The work plan for this 16-month long program was performed in three phases:

PHASE 1 - INITIAL SCREENING OF 5 CANDIDATE ALLOYS
PHASE 2 - EVALUATION OF 4 OPTIMIZED AOD HEATS
PHASE 3 - WELDABILITY TESTING OF 4 OPTIMIZED AOD HEATS

During the course of this research, progress reports were presented to the SP-7 Panel during the meeting on March 28, 1990 in Bath, Maine and again on September 19, 1990 at Mare Island Naval Shipyard. This final report covers all of the specified tasks as well as many additional studies (perfo~ed at no additional cost to the project) during the specified contract period of 16 months ending March 7, 1991.

Phase 1 - Initial Screeninq of Candidate A11ovs.

Five 200 Lb. air melt induction heats were produced to the aim compositions shown in Table 2. The furnace was lined with a rammed 90% tabular alumina refractory. The heats were tapped at a temperature of 3000° F. into shank ladles lined with Foseco No. 514 Kaltec one-piece inserts. Deoxidation additions consisted of 0.1 Lbs. 30% calcium-silicon, 0.26 Lbs. 35% zirconium-silicon and 0.1 Lbs. aluminum. The CaSi and the ZrSi were added to the furnace one minute prior to tapping. The aluminum addition was made by placing it on the bottom of the ladle and tapping onto it. The metal was held in the ladle until the temperature reached 2880” F and was then poured into sodium silicate-bonded zircon sand molds to form two 6" X 6" X 6" test blocks.

The composition of each heat was determined by optical emission spectrometry. The carbon, sulfur, and nitrogen levels were determined by combustion methods.

TABLE 2

AIM COMPOSITIONS OF AIR-MELT INDUCTION HEATS
(Wt. Percent)

	C	Mn	Si	Cr	Ni	Mo	Cu	Cb	V
HEAT 1089	0.05	0.85	0.50	1.50	4.50	0.40	-	-	-
HEAT 1090	0.05	0.85	0.50	1.50	5.50	0.40	-	-	-
HEAT 1091	0.05	0.80	0.30	0.55	5.50	0.50	-	-	0.08
HEAT 1092	0.05	0.85	0.50	1.50	5.50	0.40	-	-	0.08
HEAT 1093	0.06	0.50	0.35	0.75	1.00	0.20	1.50	0.40	-

Heate 1089 and 1090 were low carbon, higher nickel variations of HY-80 . Heat 1091 was a low carbon variation of HY-130. Heat 1092 was the same as Heat 1090 except with the vanadium addition of Heat 1091. The composition of Alloy 1093 was simply that of HSLA-80. Previous work (Ref. 4) suggested that this composition would not give adequate low-temperature toughness, but it was felt important to verify this.

All of the test blocks were normalized at 1800° F and tempered at 1250° F. The gates and risers were removed by oxy-fuel cutting using a preheat of 400° F. Several 1" cubes were saw cut from one of the test blocks cast in Heats 1089, 1090, 1091, and 1092, These samples were hardened by austenitizing at 1700°F for 2 hours followed by water quenching. The cubes were then tempered for 2 hours at temperatures ranging from 800° F to 1400° F followed by water quenching and testing for Brinell hardness. Since the measured values indicated that none of the tempering treatments employed would produce sufficiently low hardness valuee, additional samples were subjected to a double temper consisting of 1250° F, (2 hours) water quench; 1100° F, (2 hours) water quench. Since this produced the a hardness levels which would give approximately the desired strength level, this treatment was applied to the remaining test material from Heats 1089, -1090, 1091, and 1092 using 4 hour hold times at each temperature. The test blocks from Heat 1093 were austenitized at 1700°F for 4 hours followed by water quenching and then tempered at 1100°F for 4 hours followed by water quenching.

One standard 0.505 in. diameter tensile specimen and ten Charpy V-notch specimens were machined from one of the test blocks from each heat. The specimens were removed from material located at least 1 in fromthe cast surfaces. All snecimens were tested in accordance with the requirements of Ailll A370. The impact specimens were tested at temperatures of 72° F, 0° F, and -100° F. Based on the results of these teets, a series of alloys was selected for further study in Phase 2.

The purpose of the air induction melting was to quickly and economically eliminate any of the candidate alloys which would stand little chance of giving the desired mechanical properties. It was recognized that air-melted induction heats are generally not capable of matching the toughness and quality of AOD heats, but it was expected that meaningful comparisons could be drawn between the alloys.

Phase 2 - Evaluation of 4 Optimized AOD Eeats.

Using the results from Phase 1, a number of compositions were devised for further study (Table 3). Since Heat 1090 appeared to give the best toughness and since it appeared that the vanadium-containing heats did not give any particular advantages, the heats for Phase 2 were based on the composition of Heat 1090. For the heats in Phase 2, minor variations in composition were made to try to establish suitable composition ranges. In this regard, the greatest concern was the effect of carbon level on mechanical properties and weldability.

Four 2000 Lb. AOD-refined heats were produced.to the aim compositions shown in Table 3. Initial melting was done in air using induction furnaces lined with 90% tabular alumina. The heats were transferred to a l-ton AOD vessel lined with chrome-magnesite brick. The heats were transferred with carbon levels of approximately 0.30%. One hundred pounds of lime were added to the veesel and the heat was blown with argon and oxygen to approximately the O% carbon level. The slag was poured off to remove the phosphorus and was rebuilt using 100 Lbs of lime and adequate aluminum to thin the slag and to act as fuel for further processing. After making the final alloying additions, and stirring with argon, the heats were tapped at 3000°F. into a ladle lined with a castable 90% tabular alumina refractory. The deoxidation consisted of 1 Lb. 30% calcium silicon, 2.6 Lbs. zirconium silicon and 1 Lb. aluminum, all added to the tapping stream. The metal was held in the ladle until the temperature decreased to 2860° F. and was then poured into sodium silicate-bonded zircon sand molds to produce test blocks measuring 12" X 1211 x 2211 (30.5 cm X 30.5 cm X 56 cm). The composition of each heat was determined as described previously.

TABLE 3

AIM COMPOSITIONS OF 2,000 LB AO13 HEATS
(Wt. Percent)

	C	Mn	S,	Cr	Ni	Mo
HEAT 1094	0.04	0.85	0.35	1.50	5.50	0.45
HEAT 1103	0.02	0.65	0.30	1.50	5.50	0.45
HEAT 1118	0.06	0.85	0.50	1.50	5.50	0.45
HEAT 1132	0.04	0.90	0.45	1.35	5.50	0.60

The test blocks were normalized and tempered prior to riser removal by oxy-fuel cutting. They were then hardened and tempered using the results of Phase 1 as a guide for specific temperature selection. In several cases, it was necessary to re-heat treat to achieve the hardness levels felt to be appropriate. The specific heat treatments qiven to the individual blocks are show-in Table 4 below:

TABLE 4

HEAT TREATMENT OF AOD TEST BLOCKS

	NOPMALIZE AND TEMPER (°F)	HARDENING TEMPERATURE (°F)	DOUBLE TEMPER (°F)
HEAT 1094	1800 AC 1250 WQ	1700 WQ	1275 WQ 1100 WQ
HEAT 1103	1800 AC 1250 WQ	1700 WQ	1200 WQ 1100 WQ
HEAT 1118	1800 AC 1250 WQ	1700 WQ	1275 WQ 1125 WQ
HEAT 1132	1800 AC 1250 WQ	1700 WQ	1250 WQ 1100 WQ

Standard tensile and impact specimens were removed from the test blocks according to the requirments of MIL-STD-23008 Revisions B, C, and D. For Revisions B and C, the specimens were located a minimum of 1 in from any surface of the block. For Revision D, the specimens were taken from the 1/4 T location. Again, the specimens were tested in accordance with ASTM A370. The impact specimen test temperatures 72° F, 0°F, and -100°F.

In order to better understand the response of the alloys to heat treatment and also the mechanical property behavior, specimens from Heat 1118 were subjected to heat treating studies using a Gleeble Model 1500. The specimens were austenitized at 1700°F for 5 minutes and then subjected to continuous cooling rates ranging from 1.5 to 4182°F/minute. The dilatation of the specimens was monitored to detect the onset and completion of transformations.

The microstructure of Heat 1132 was also examined by optical metallography.

Phase 3 - Weldability Testina of AOD Heats.

Weldability of each AOD heat was evaluated using the Controlled Thermal Severity (CTS) Test and the Tekken Test. For comparison, tests were also conducted with cast HY-80. Test specimens were machined to the dimensions shown in Figures 4 and 5. In all cases, the specimen thickness was 1 in.

All welding was performed by GMAW and FCAW using a controlled heat input of 55 kJ/in (2.2 Kj/mm). Specific welding variables used in all test welds are presented in Table 5.

TABLE 5
WELDING PARAMETERS FOR GMAW AN13..FCAW
CRACK SUSCEPTIBILITY TESTS

	GMAW	FCAW
CURRENT (DCEP)	340 AMPS	340 ANPS
VOLTAGE	30 VOLTS	30 VOLTS
TRAVEL SPEED	11 1PM	11 1PM
SHIELDING GAS	98% Ar + 2% 0_2	75% Ar + 25% $C0_2$
FILLER NATERIAL	MIL-1OOS-1	MIL-101TM
FILLER DIAMETER	1/16 in.	1/16 in.

All welding of the experimental alloys and the reference HY-80 was performed without preheating. The specimens were allowed to remain at room temperature for 72 hours after welding. Three cross sections were then cut from each specimen. These were ground and then examined for cracking. Microhardness surveys (Knoop, 500 g) were also performed on cross sections from the Tekken tests.

Diffusible hydrogen was measured for both GMAW and FCAW weld metal deposits using gas chromatography and also the glycerine method. The gas chromatography was used in accordance with AWS B4.3. The glycerine method is only standardized in Japan but is also used in this study because of its simplicity.

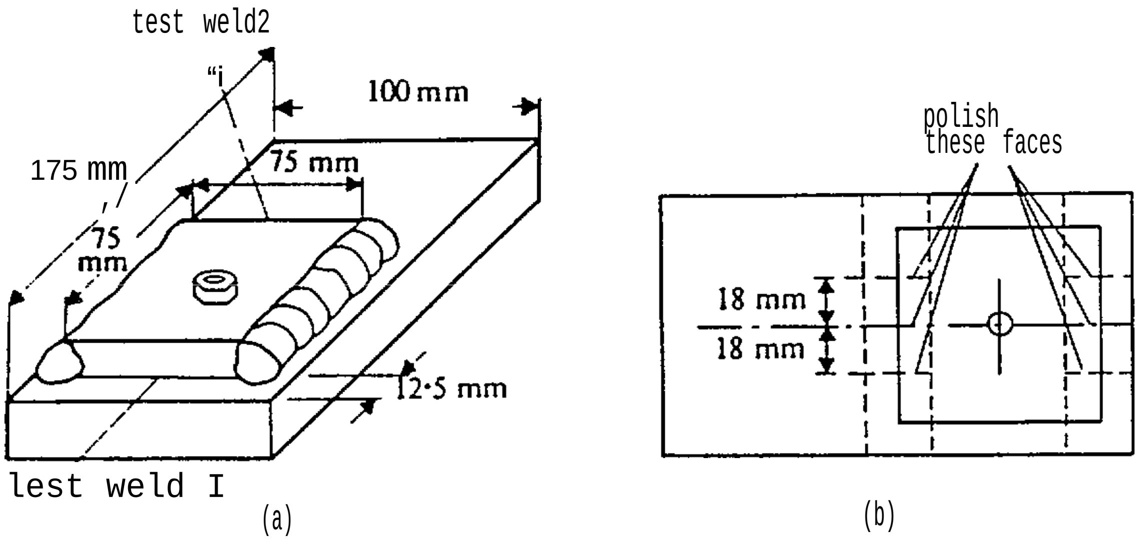


Figure 4.

Configuration of Controlled Thermal Severity (CTS) test specimens.

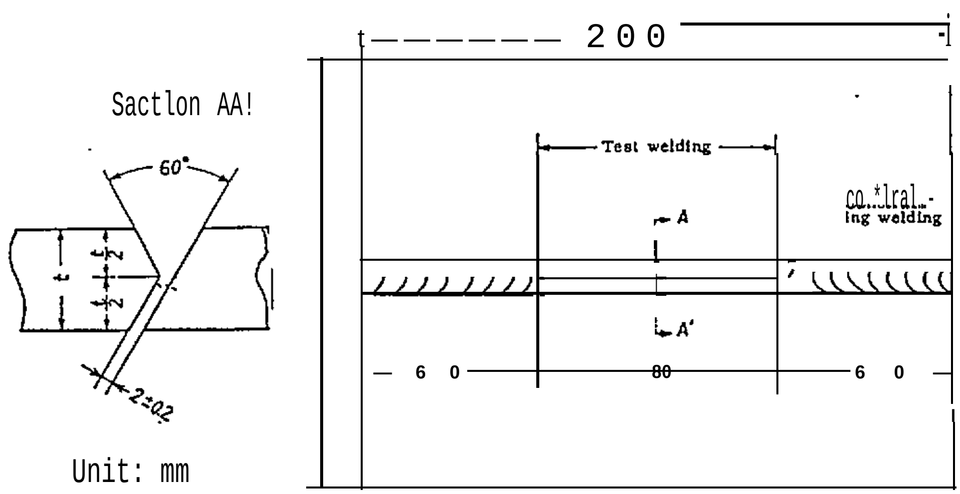


Figure 5.

Configuration of Tekken test epecimens.
(From Japanese Industrial Standard JIS Z 3158 (1983).)

RESULTS

PHASE 1 - Initial Screening of 5 Candidate Alloys

The actual compositions of the air melt induction heats are given in Table 6.

TABLE 6

ACTUAL COMPOSITIONS OF AIR MELT INDUCTION HEATS
(Wt. Percent)

	HEAT 1089	HEAT 1090	HEAT 1091	HEAT 1092	HEAT 1093
C	0.059	0.041	0.042	0.048	0.050
Mn	0.96	0.90	0.84	0.79	0.51
Si	0.60	0.56	0.59	0.46	0.35
Cr	1.59	1.65	0.61	1.46	0.82
Ni	4.52	5.77	5.81	5.45	1.01
Mo	0.41	0.43	0.49	0.39	0.20
Cu	0.10	0.11	0.10	0.10	1.82
Cb	0.01	0.01	0.01	0.01	0.05
V	0.02	0.02	0.02	0.08	0.01
S	0.014	0.014	0.011	0.012	0.008
P	0.011	0.011	0.008	0.010	0.010
Al	0.045	0.069	0.052	0.049	0.047
Zr	0.037	0.012	0.035	0.024	0.020
N	174 ppm	103 ppm	97 ppm	132 ppm	119 ppm

Tempering Response

The response of the air-melt induction heats to tempering is illustrated in Figures 6 through 10. Heats 1089, 1090, 1091 and 1092 all behaved in a similar manner. The as-quenched hardness was approximately 34 Rc and this hardness level decreased only slightly with tempering temperatures up to 1000°F. Tenqiering temperatures between 1000°F and 1200°F resulted in sharply decreasing hardness values. Tempering above 1200°F caused the hardness values to increase markedly.

Heat 1093 gave a much different response to tempering. Tempering temperatures up to 800°F did not change the hardness. Tempering between 800°F and 900°F significantly increaeed the hardness. Temperatures above 900°F caused hardness levels to decrease again.

Mechanical Properties

The measured hardness values, the tensile properties and the impaCt toughness values are given in Tables 7, 8, and 9 respectively.

TABLE 7

HARDNESS VALUES OF AIR-MELT INDUCTION HEATS

HEAT NUMBER	BRINELL HARDNESS (3000 Kg)
1089	229
1090	229
1091	229
1092	229
1093	207

TABLE 8

TENSILE PROPERTIES OF AIR MELT INDUCTION HEATS

	HEAT 1089	HEAT 1090	HEAT 1091	HEAT 1092	HEAT 1093
UTS (KSI)	99.2	96.4	104.4	110.8	90.7
S_Y (KSI)	76.7	76.6	85.7	90.5	79.2
% E	14	20	24	22	27
% R. A.	22	48	67	65	70

TABLE 9

IMPACT TOUGHNESS OF AIR-MELT INDUCTION HEATS

HEAT NUMBER	TEST TEMPERATURE (°F)	IMPACT TOUGHNESS (Ft Lbs)	AVERAGE (Ft Lbs)
1089	72	108, 106	106.5
	0	99, 75, 100, 81	88.75
	-100	58, 59, 58, 59	58.5
1090	72	125, 119	122
	0	99, 99, 108, 109	103.75
	-100	59, 79, 78, 69	71.25
1091	72	87, 110	98.5
	0	94, 96, 96, 68	88.5
	-100	75, 70, 73,. 6 8	71.5
1092	72	97, 93	94.75
	0	89, 87, 87, 76	84.75
	-100	85, 80, 81, 79	81.25
1093	72	118, 117	117.5
	0	93, 101, 102, 88	96
	-1oo	8, 7, 6, 6	6.75

Hardness, tensile and CVN impact toughness data for these alloys are shown in Figures 11 through 17.

During preparation of the teSt specimens, it was found that the test blocks from Heats 1089 and 1090 contained numerous hydrogen flakes (internal hydrogen cracks). The tensile test results for these two heats were affected by these pre-existing flaws. Had these defects not been present, it is likely that higher yield strengths and better ductility would have been measured for Heats 1089 and 1090. Examination of the fractures of the impact specimens indicated that the hydrogen flakes did not influence the measured toughness values.

PHASE 2 - Production of 4 Optimized AOD heats

The actual compositions of the four AOD heats of Phase 2 are shown in Table 10.

TABLE 10

ACTUAL COMPOSITIONS OF 2,000 LB AOD HEATS
(Wt. Percent)

	HEAT 1094	HEAT 1103	HEAT 1118	HEAT 1132
C	0.044	0.032	0.055	0.033
Mn	0.82	0.50	0.60	0.99
Si	0.46	0.33	0.22	0.41
Cr	1.67	1.70	1.75	1.38
Ni	5.48	5.32	5.27	5.41
Mo	0.47	0.46	0.48	0.56
S	0.003	0.007	0.010	0.007
P	0.011	0.011	0.008	0.004
Al	0.028	0.010	0.028	0.014
Zr	0.038	0.034	0.044	0.014
N	53 ppm	44 ppm	39 ppm	45 ppm

The hardness values, the tensile properties and the impact toughness data obtained from the AOD heats are presented in Tables 11, 12 and 13 respectively.

TABLE 11

HARDNESS OF AOD HEATS

HEAT NUMBER	LOCATION	BRINELL HARDNESS (3000 Kg)
1094	1 in.	241
1103	1 in.	217
	1/4 T	217
1118	1 in.	207
	1/4 T	207
1132	1 in.	235
	1/4 T	235

TABLE 12

TENSILE PROPERTIES OF AOD HEATS

	LOCATION	UTS (KSI)	(KSI)	% E	% R. A.
HEAT 1094	1 in.	107.4	91.0	22	60
	1/4 T	106.9	90.6	20	58
HEAT 1103	1 in.	97.0	81.5	22	57
	1/4 T	98.1	80.7	21	51
HEAT 1118	1 in.	100.6	68.7	21	61
	1/4 T	99.6	68.9	20	64
HEAT 1132	1 in.	104.2	82.4	19	52
	1/4 T	103.5	84.1	21	52

TABLE 13

IMPACT TOUGHNESS OF AOD HEATS

HEAT NUMBER	LOCATION	TEST TEMPERATURE (°F)	IMPACT TOUGHN!JSS (Ft Lbs)	AVERAGE (Ft Lbs)
1094	1 in.	72	S8, 79, 78	81.6
		0	80, 76, 74	76.6
		-100	55, 49, 56	53.3
	1/4 T	72	90, S9, 93	90.6
		0	89, 8S, S8	88.3
		-100	52, 46, 45	47.6
1103	1 in.	72	103, 76, 71	85.3
		0	80, 75, 70	75
		-100	55, 45, 62	54
	1/4 T	72	98, 101, 102	100.3
		0	100, 89, 115	101.3
		-100	61, 50, 68	59.6
1118	1 in.	72	79, 70, 74, 91	78.5
		0	72, 67, 60	66.3
		-100	6S, 71, 66, 65	67.5
	1/4 T	72	100, 102, 104, 10s	103.5
		0	120, 109, 100	109.6
		-100	90, 72, 93, 104	89.8
1132	1 in.	72	72, 73, 77	74
		0	67, 69, 76, 77	72.2
		-100	56, 60, 61	59
	1/4 T	72	104, 108, 122	111.3
		0	88, 90, 98	92
		-100	46, 52, 55, 57	52.5

The measured hardness values the yield strenqths and the impact toughness data from the AOD heats are presented graphically in Figures 18 through 23.

Metallographv

The microstructure of Heat 1132 appeared to consist of a mixture of tempered martensite and lower bainite (Fig. 24). It should be recognized, however, that these transformation products are difficult to distinguish optically in low-carbon steels which have been tempered at high temperatures.

Thermal Analvses durinq Continuous Coolinq

Based on dilatometry tests of Eeat 1118, the phase transformations were plotted for different cooling rates to produce the Continuous Cooling Transformation (CCT) diagram presented in Figure 25. It is clear that these low carbon steels generally will produce martensite and~or bainite over a very wide range of cooling rates.

PHASE 3 - Weldability Testinq of 4 Optimized AOD Heats

The results of the weldability tests are presented in Table 14. All of the heats of the experimental alloys passed the CTS tests for both FCAW and GMAW. These materials also survived the GMAW Tekken tests without cracking. In addition, Heats 1103 and 1132 exhibited no cracking in the FCAW Tekken tests. Some cracking was found in the FCAW Tekken tests of Heats 1094 and 1118. However, the cracks all appeared to be confined to the weld metal.

Severe HAZ cracking was observed in both the GMAW and FCAW Tekken tests of the cast HY-80 material. HAZ cracking was also found in the FCAW CTS test. Only in the GMAW CTS test was no cracking observed for this material.

TABLE 14

TEKKEN AND CTS TESTS PERFORMED BY GMAW & FCAW AT 55 KJ/IN WITHOUT PREHEATING.

TEST TYPE	HEAT NUMBER	GNAW PROCESS	FCAW PROCESS
TEKKEN	1094	NO CRACKS	CRACKS IN WELD METAL
	1103	NO CRACKS	NO CRACKS
	1118	NO CRACKS	CRACKS IN WELD METAL
	1132	NO CRACKS	NO CRACKS
	HY-80	CRACKS IN HAz	CIU+CKS IN HAz
CTS	1094	NO CRACKS	NO CRACKS
	1103	NO CRACKS	NO CRACKS
	1118	NO CEACKS	No CRACKS
	1132	NO CRACKS	NO CRACKS
	HY-80	NO CRACKS	CRACKS IN HAz

The differing locations of the craking eXhibited by the experimental alloys and by HY-80 in the Tekken tests is illustrated in Figs. 26 and 27.

Weld Metal Toughness

The CVN impact toughness of weld metal deposited on Alloy 1132 was evaluated and the results are given in Table 15.

TABLE 15

CVN IMPACT TOUGHNESS OF WELD METAL DEPOSITED BY GNAW AND FCAW USING CONDITIONS GIVEN IN TABLE 4

TEST TEMPERATURE (°F)	IMPACT TOUGHNESS (Ft Lbs)
72	104, 100
0	88, 90 100
-60	50, 47, 48

Since the Naval requirements for weld metal include CVN impact toughness values of 35 ft-lbs at -60 and 60 ft-lbs at 0, the welds in alloy 1132 exceed the present Navy requirements for weld metal. It is anticipated that the welds in the other experimental alloys would yield similar results.

Hvdrogen Content of Welds

Diffusible hydrogen content was determined for welds deposited by both GMAW and FCAW. From the Table 16, it can be seen that the diffusible hydroqen content of the welds denosited bv FCAW were greater thanthat deposited by GMAW.

TABLE 16

DIFFUSIBLE HYDROGEN TEST USING GAS CHROMATOGRAPHY PER AWS 4.3 AND GLYCERINE (FOR COMPARISON)

	($ml\ H_2/100\ g$)	($ml\ H_2/100\ g$)
MIL-100S-1	1.4	0.8 1.0
MIL-101TM	6.4	3.5 4.0

Hardness Profile of Weld Joints

Microhardness profiles were obtained from the Tekken test welds of each material. From the profilee plotted in Figures 28 through 31, the general trends are similar for each alloys 1095, 1103, 1118 and 1132. Clearly, the maximum hardness in the HAZ does not exceed 360 KHN (500 g) which is far less than that measured fOr HY-80 (Fig. 32) .

Microstructure of Weld and HAZ

The microstructure of the AOD test blocks at the 1" and quarter thickness locations were similar. Both contain low-carbon tempered martensite structures as illustrated in Figure 22 for alloy 1132. When alloy 1132 was welded by GNAW in the Tekken test and for mechanical properties evaluation, the microstructure in the weld metal was a mixture of bainite and martensite as shown in Figure 33. The excellent CVN touahness values corresponding to this microstructure are given in Table 15.

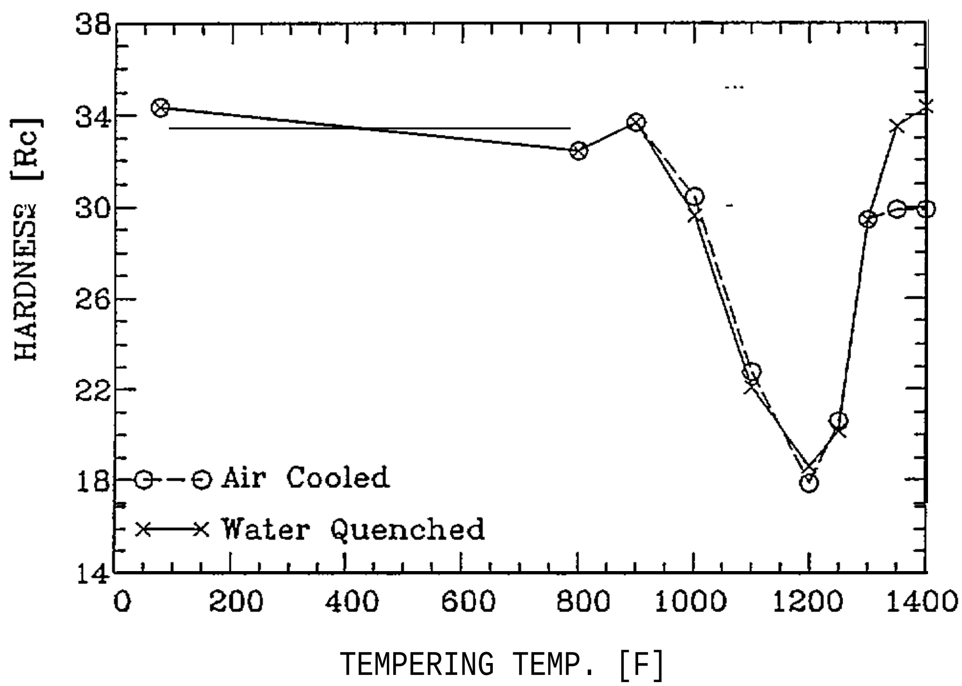


Figure 6.

Tempering response of heat 1089.

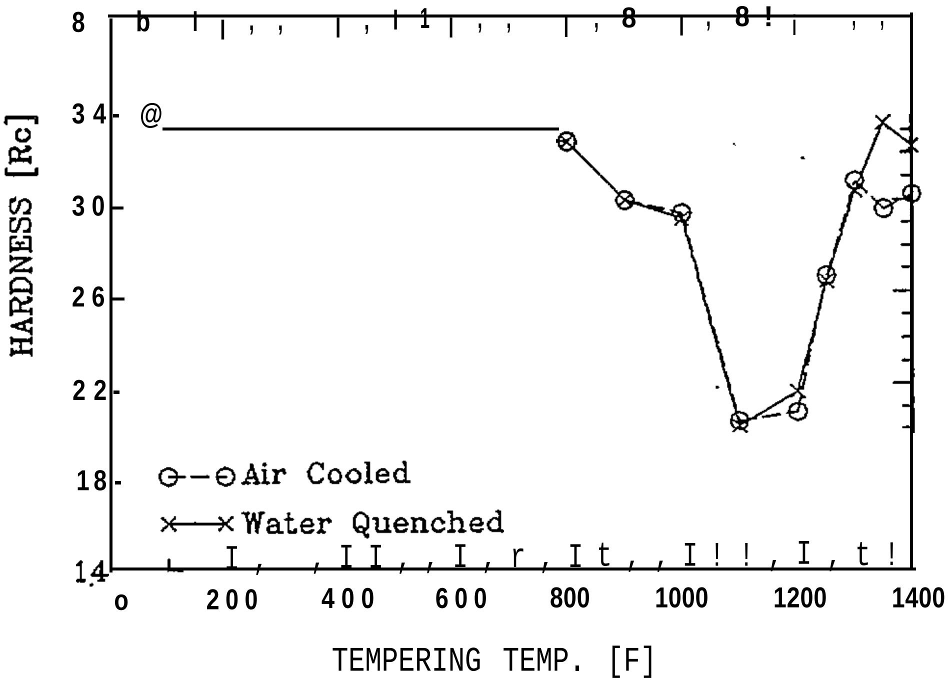


Figure 7.

Tempering response of Heat 1090.

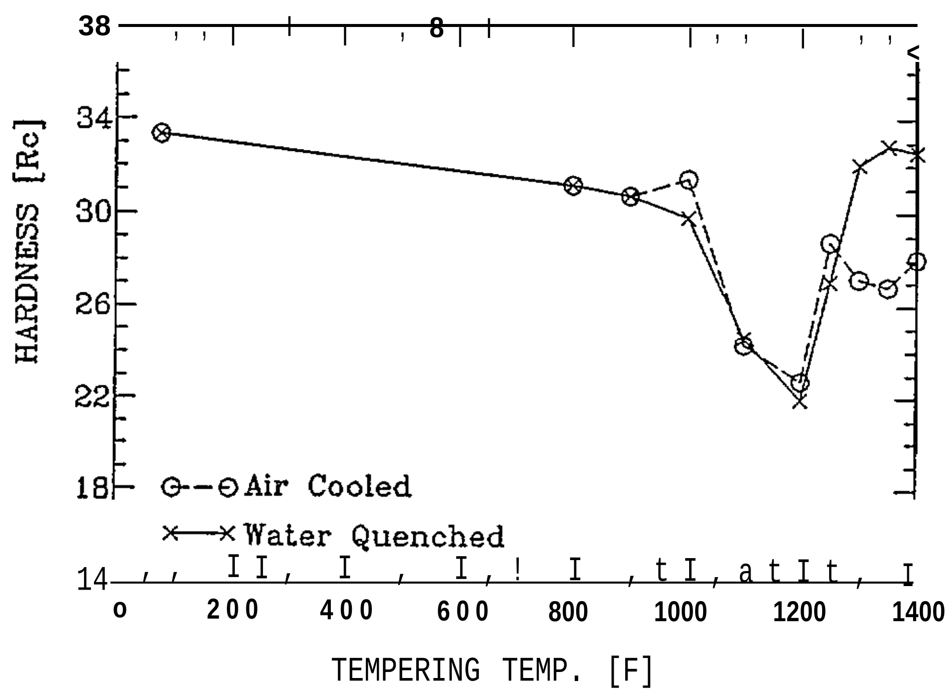


Figure 8.

Tempering response of Heat 1091.

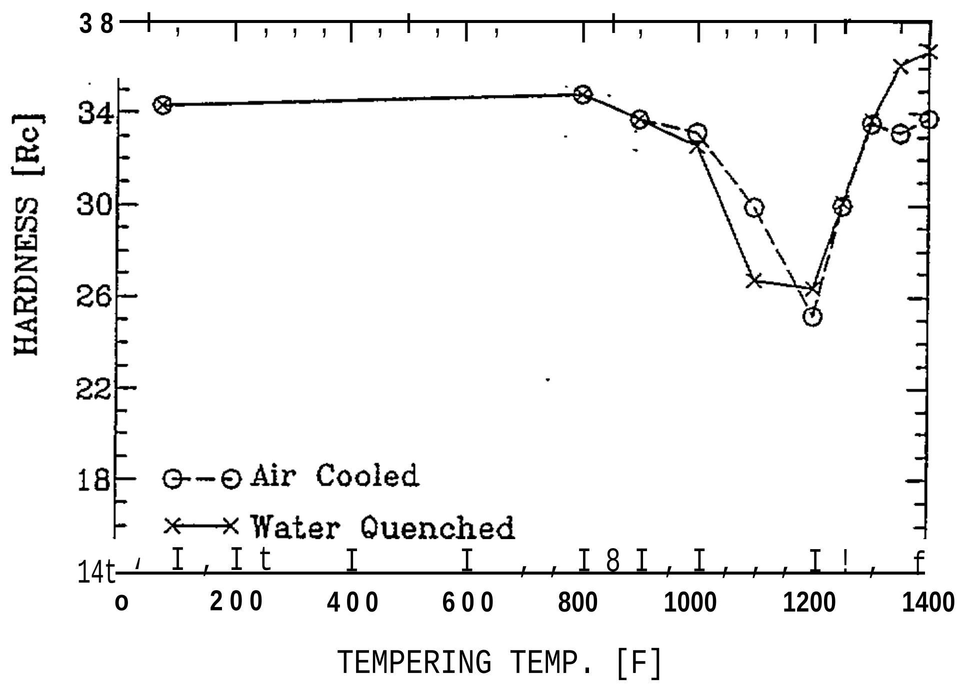


Figure 9.

Tempering response of Heat 1092.

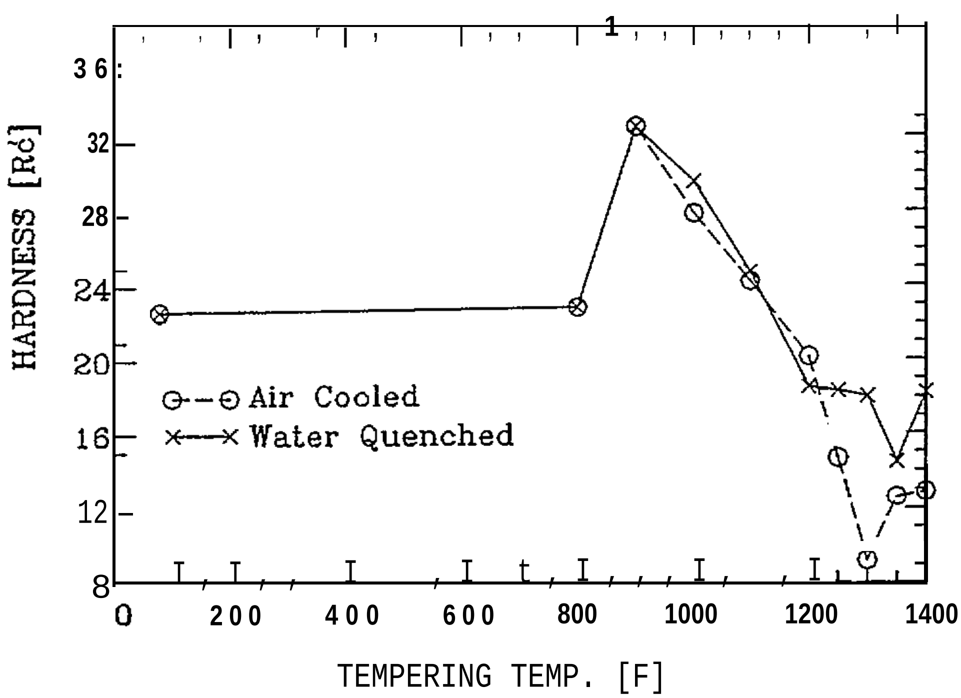


Figure 10.

Tempering response of Heat 1093.

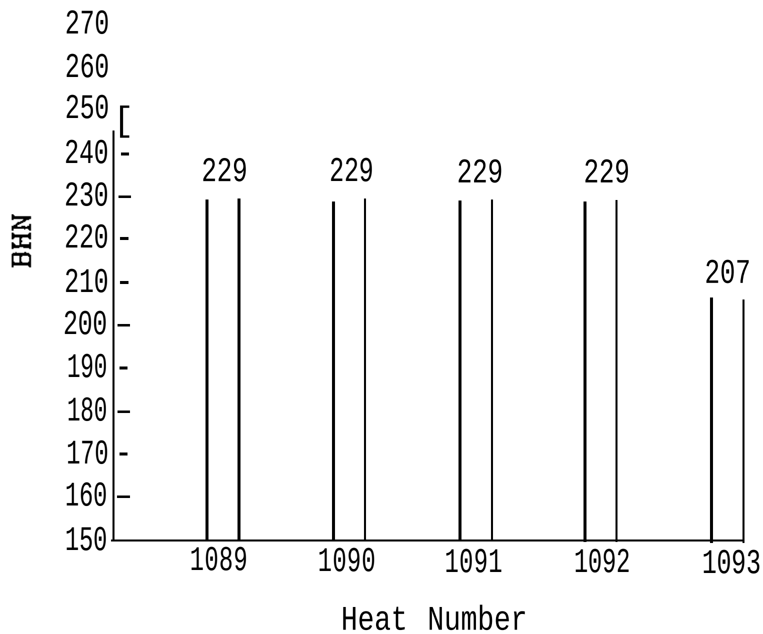


Figure 11.

Brinell hardness of air-melt induction heats.

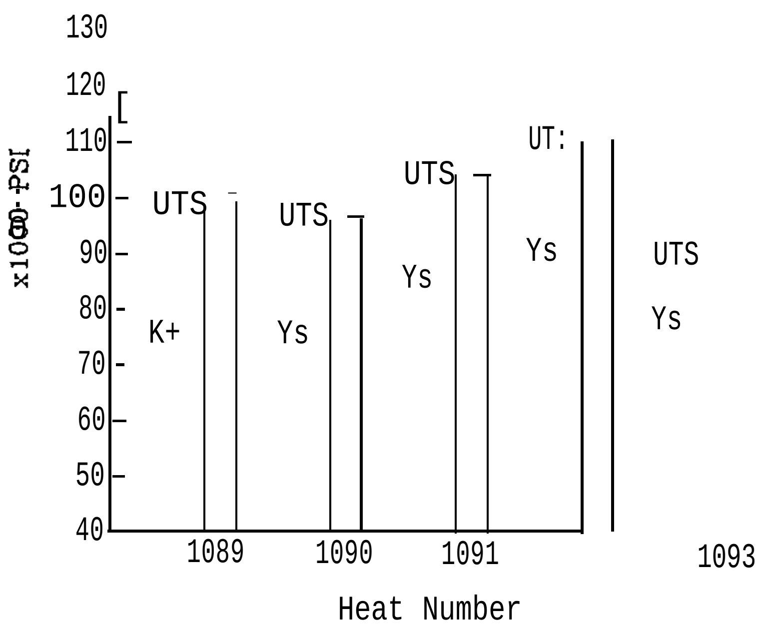


Figure 12.

Ultimate tensile and yield strengths of air-melt induction heats.

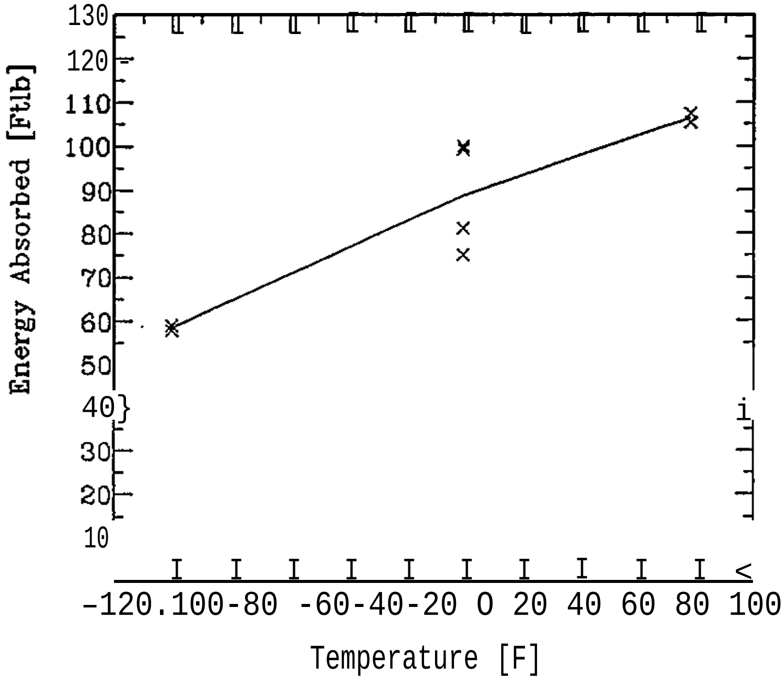


Figure 13.

Impact toughness of Heat 1089.

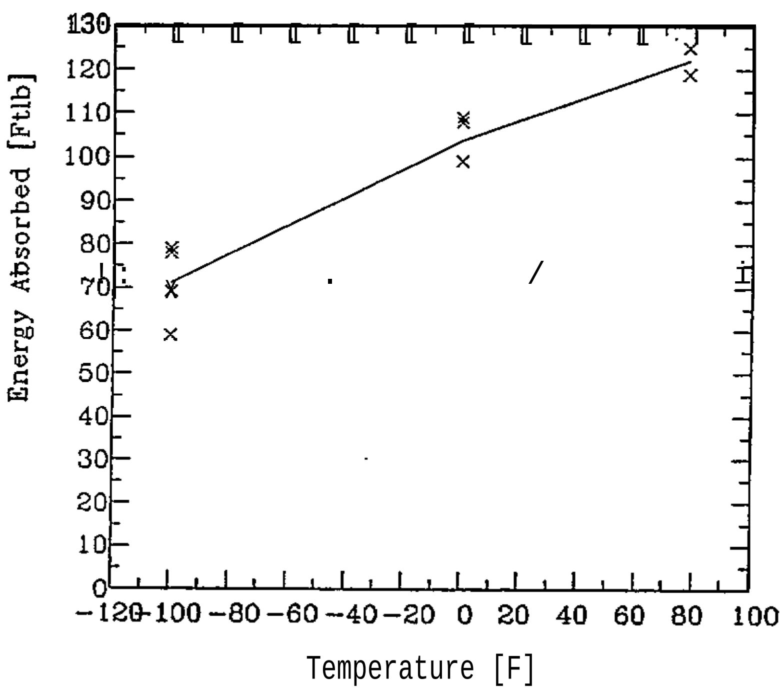


Fig. 14.

Impact toughness of Heat 1090.

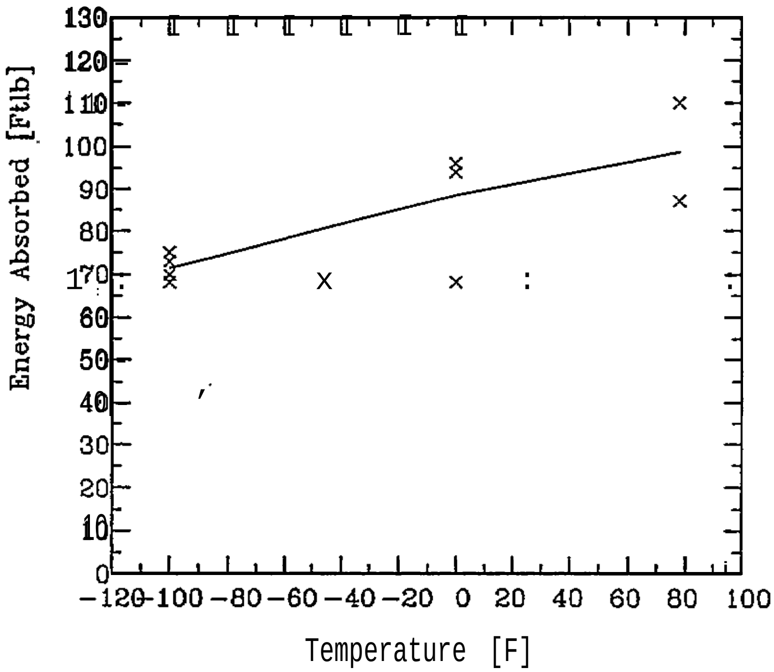


Figure 15.

Impact toughness of Heat 1091.

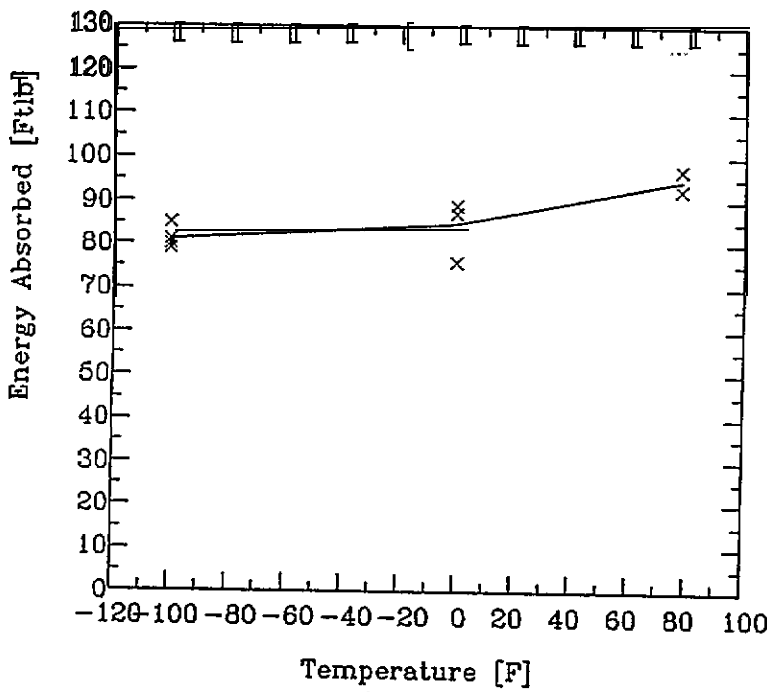


Figure 16.

Impact toughness of Heat 1092.

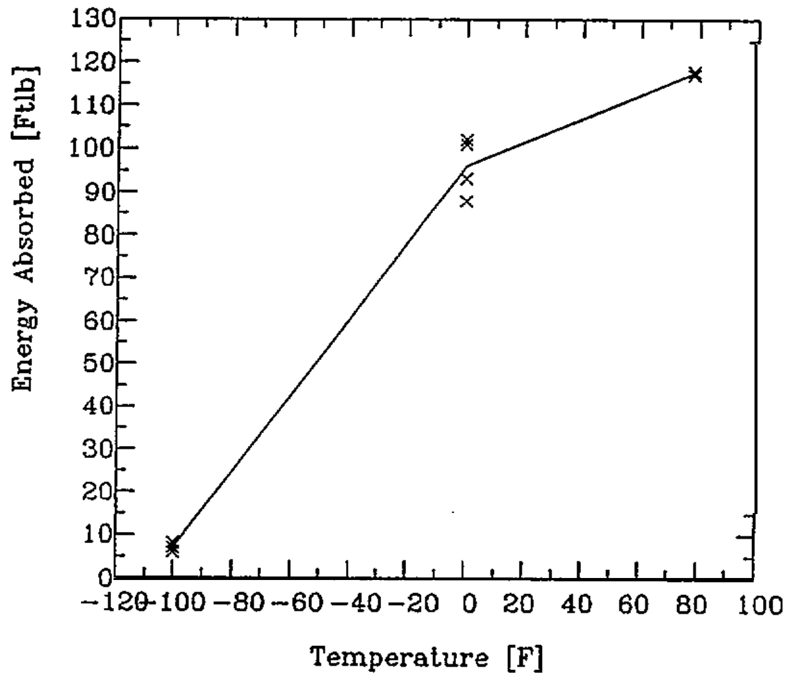


Figure 17.

Impact toughness of Heat 1093.

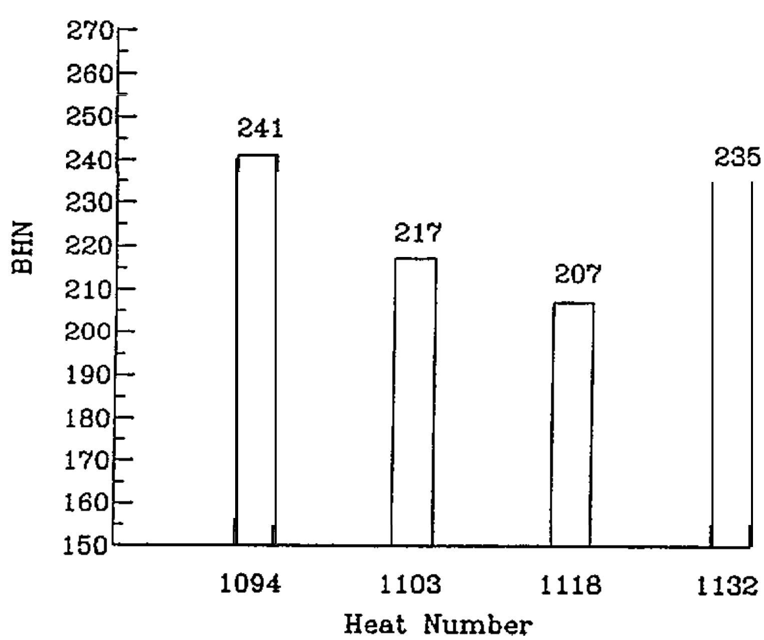


Figure 18.

Brinell Hardness of AOD heats

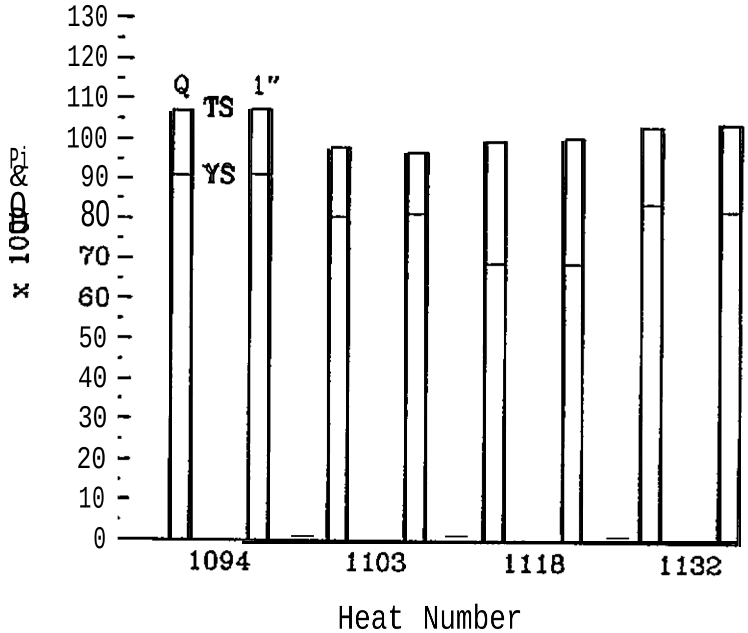


Figure 19.

Ultimate tensile and yield strengths of AOD heats.

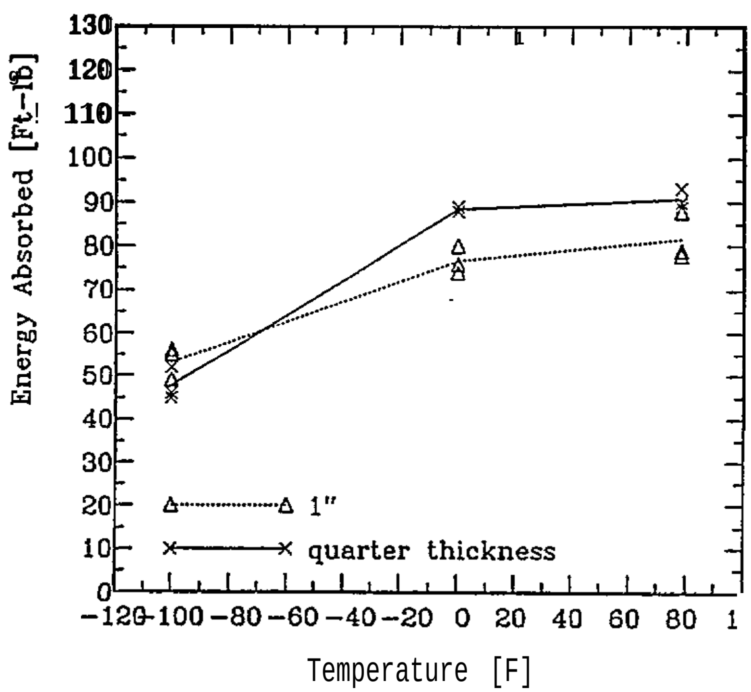


Figure 20.

Impact toughness of AOD Heat 1094.

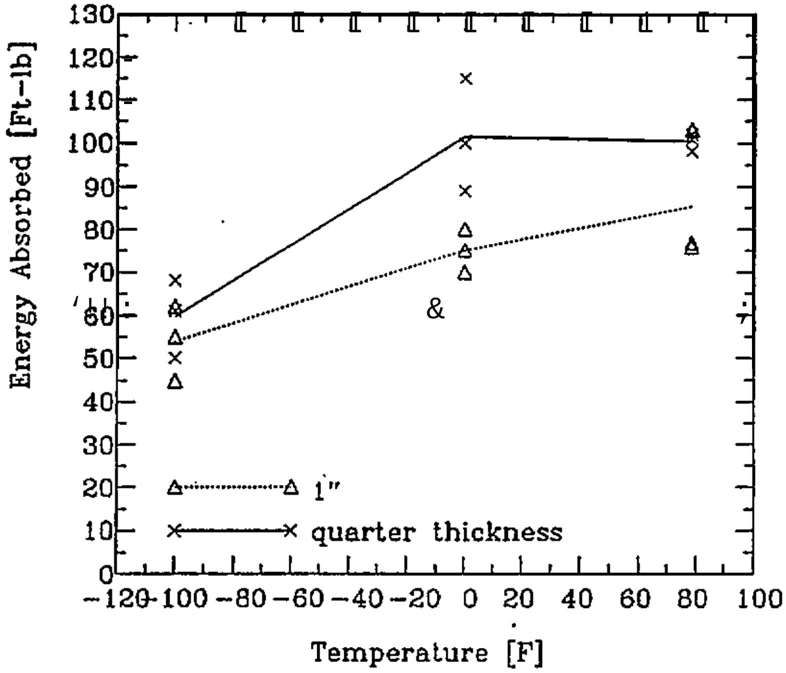


Figure 21.

Impact toughness of AOD Heat 1103.

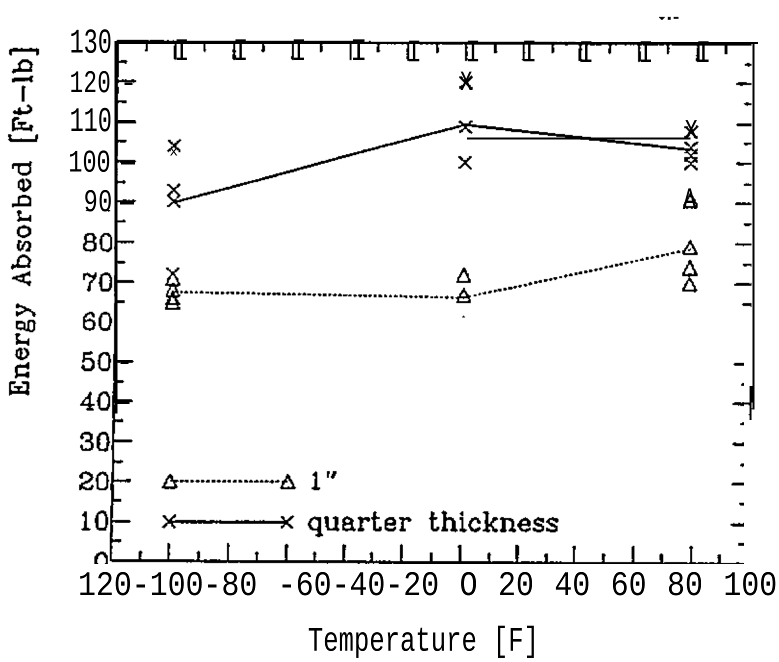


Fig. 22.

Impact toughness of AOD Heat 1118.

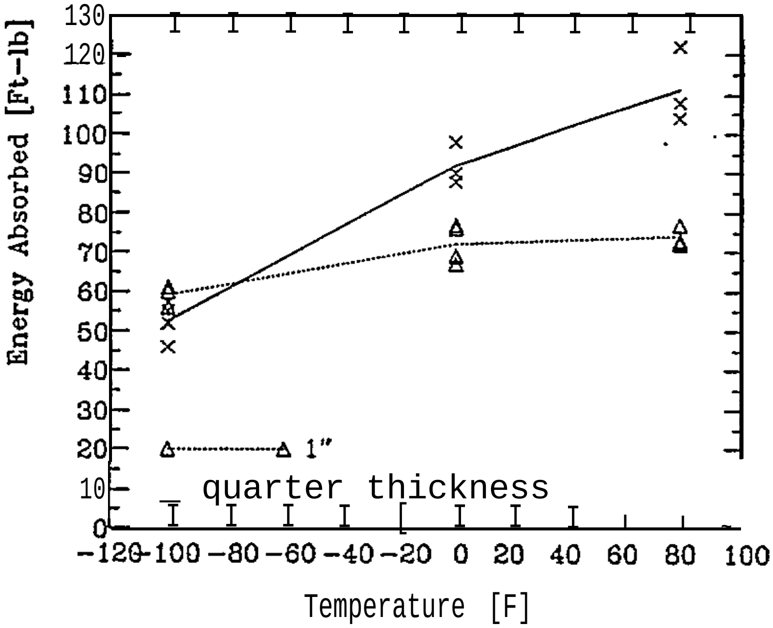


Fig. 23.

Impact toughness of AOD Heat 1132.

Figure 24.

Photomicrograph showing microstructure of Heat 1132. Tempered martensite and lower bai.nite. (2% Nital) (1000X)

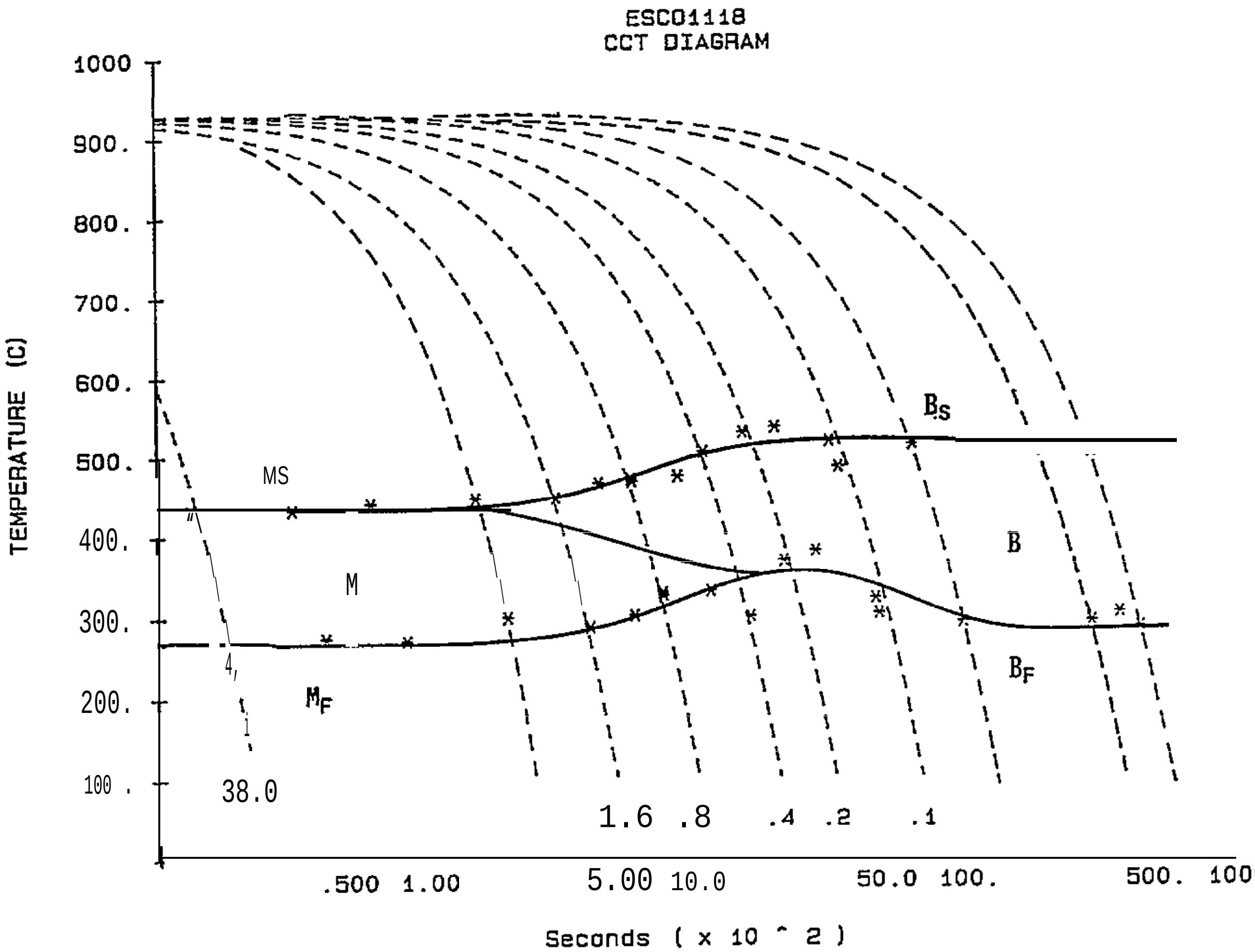


Figure 25.

continuous cooling diagram for Heat 1118.

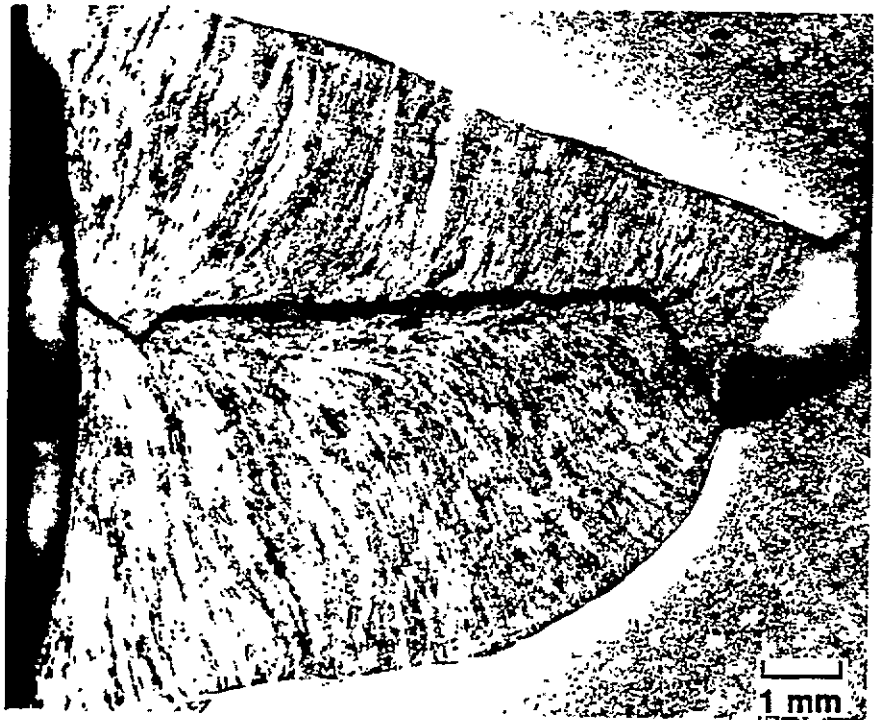


Figure 26.

Weld metal crack in FCAW Tekken test of Heat 1118.

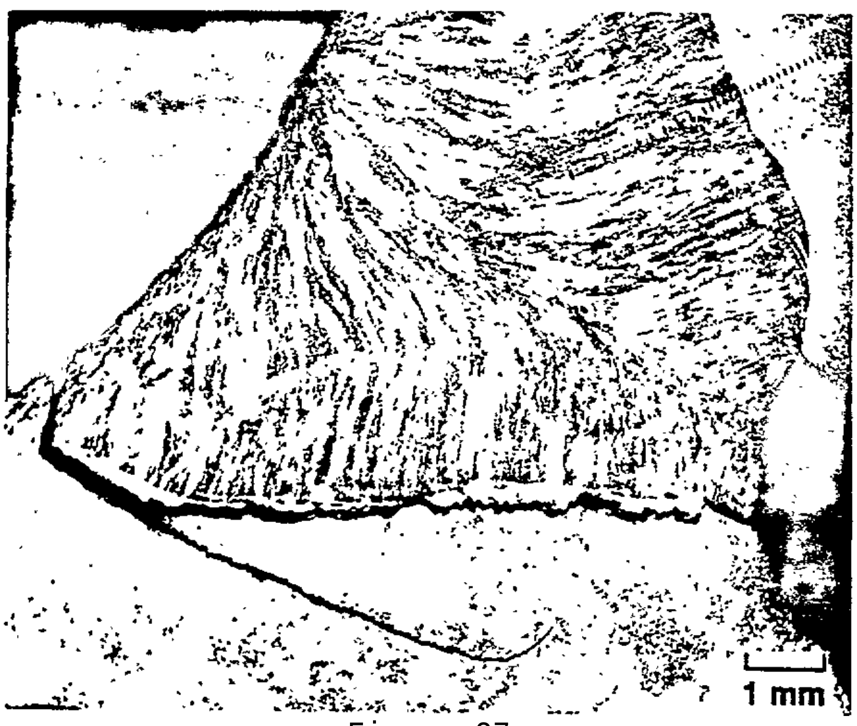


Figure 27.

HAZ crack in FCAW Tekken test of cast HY-80.

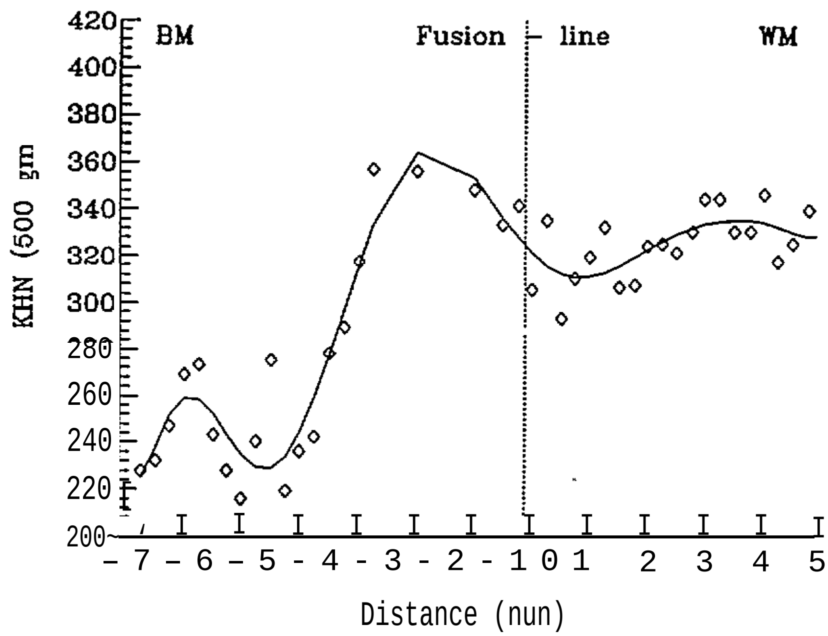


Figure 28.

Microhardness survey of Haz of GMAW Tekken Test in Heat 1094.

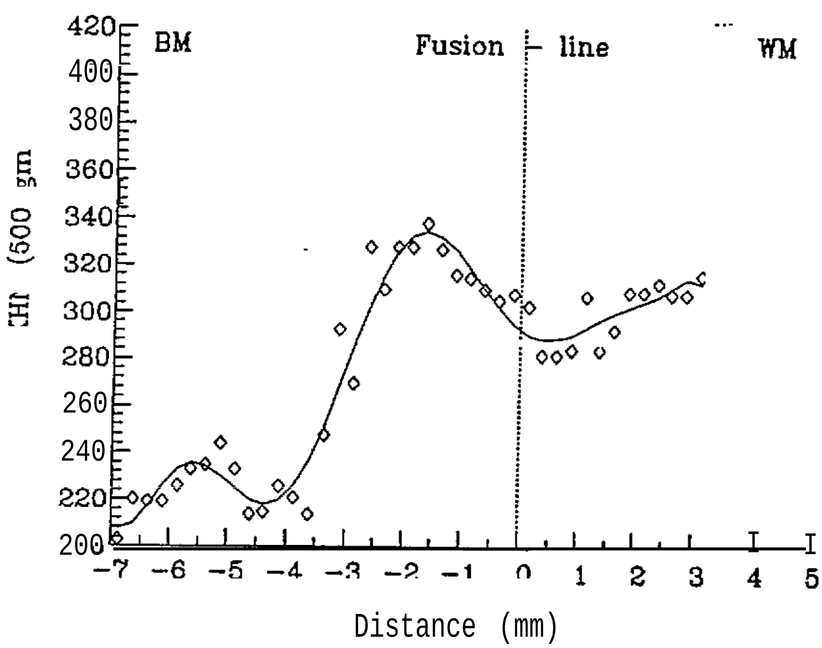


Figure 29.

Microhardness survey of HAZ of GMAW Tekken test in Heat 1103.

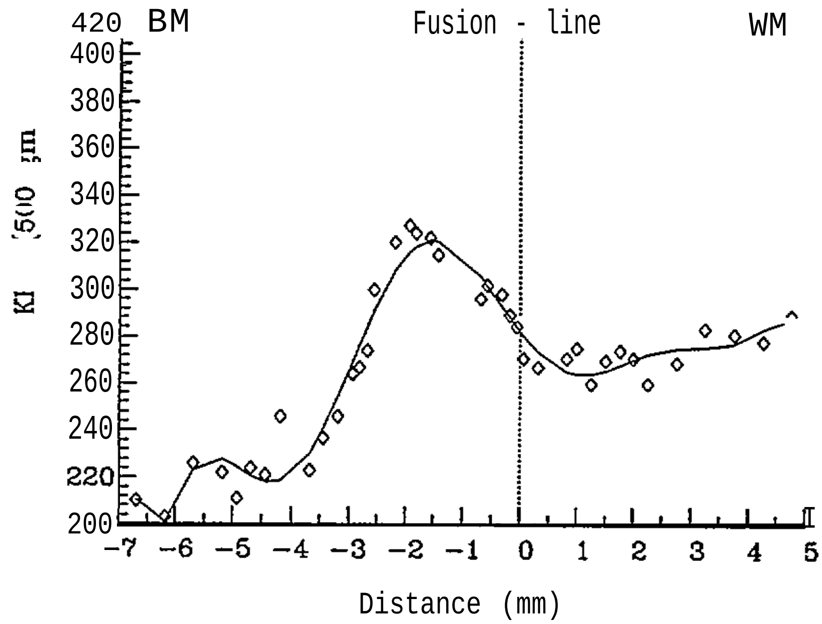


Fig. 30.

Microhardnees survey of HAZ of GMAW Tekken test in Heat 1118

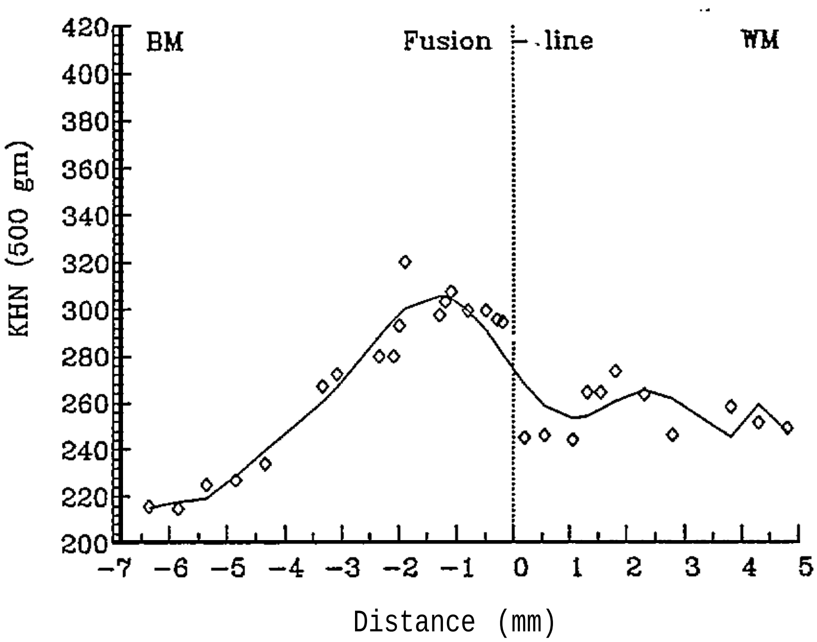


Figure 31.

Microhardness survey of Gmaw Tekken test in Heat 1132.

Weld Hardness Profile HY-80 (FCAW)

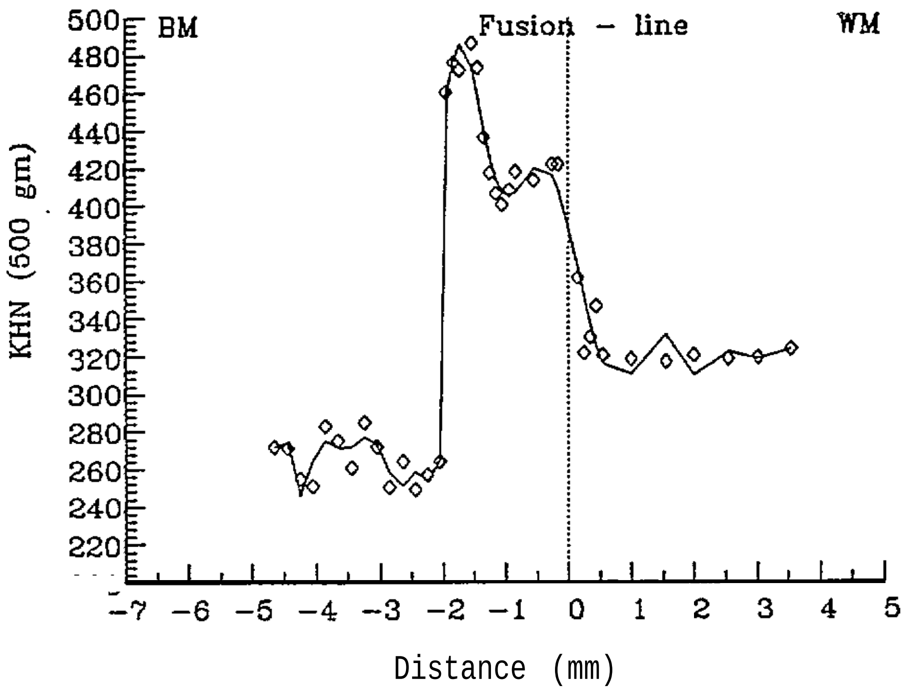


Fig. 32.

Microhardness survey of HAZ of FCAW Tekken test of HY-80.

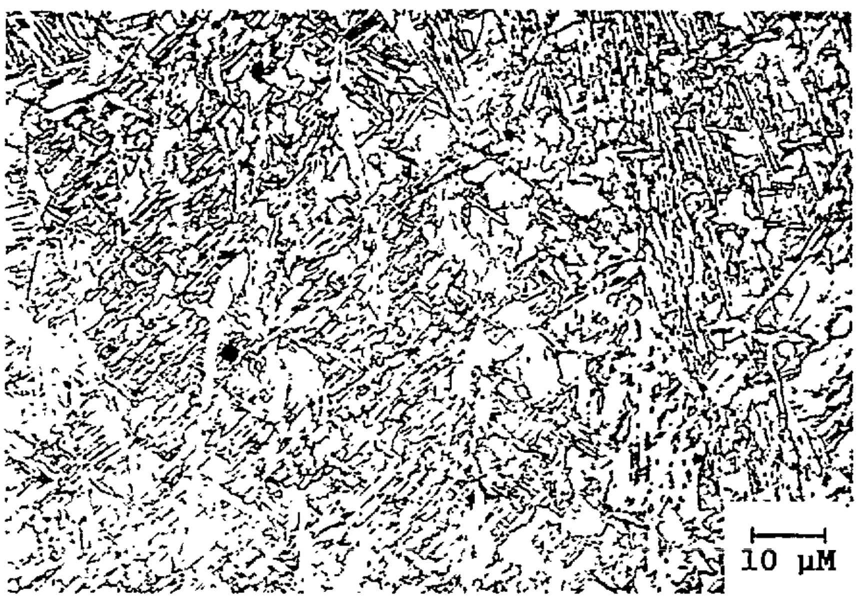


Figure 33.

Microstructure of GMAW weld metal
from Tekken test of Heat 1132.

(2% Nital) (1000X)

DISCUSSION

Examination of the mechanical property data for the air-melt induction heats reveals that all four experimental alloy compositions developed hardness levels which would be expected to give the desired level of yield strength. The measured yield strength values for the first two heats (1089 & 1090) , however, were slightly below the specified minimum value of 80 KSI. It should be pointed out, however, that in both cases, the material contained pre-existing "hydrogen flakes" (cracks) . These defects undoubtedly resulted in anomalously low measurements of yield strength. Had these defects not been present, all four heats would probably have exhibited acceptable yield strength values.

The presence of hydrogen flakes in some of the test blocks should not be alarming or even unexpected. There is no effective method of removing hydrogen when induction melting in air and so hydrogen levels are characteristically high (e.g. 6 ppm). At these hydrogen levels, <u>any</u> high strength steel casting would be expected to develop hydrogen flakes. This is one of the primary reasons that dead-melting is rarely used for producing high strength steel castings.

The measured impact toughness values for all of the air-melt heats of the experimental alloy heats were "very good, easily exceeding the specified minimum values for HY-80. Considering the fact that air-melting produces inherently "dirty" steel, heats, this is excellent performance.

The heat having the composition of wrought HSLA-80 (1093) exhibited a strength level slightly below the 80 KSI minimum specified for HY-80. While the yield strength level came close to meeting the desired value, the low temperature toughness was extremely low. This is consistent with the previous evaluations of this material in the cast form (Ref. 4). Undoubtedly, the strength could have been increased into the desired range by tempering at a slightly lower temperature. However, the previous work with this material has shown that tempering below 1100 F. results in markedly lower toughness levels. Even if the strength were increased through the use of a lower tempering temperature, the toughness would not have been improved and the material would still have been an unacceptable substitute for cast HY-80.

Three of the test blocke poured in the AOD heats (1094, 1103 and 1132) met the strength requirements specified in MIL STD 23008 Revs. B, C and D. The block poured in Heat 1118 was below the minimum etrength. Undoubtedly, the strength of this block could have been brought into the desired range with a slightly different heat treatment. With the double temper heat treatments employed for the experimental alloys, it is more difficult to predict hardnees and strength levels than with single tempering cycles commonly used with most low alloy steels. Given more experience, however, such predictions should become easier.

Examination of the impact toughness data for the AOD heats indicates that all easily passed the toughness requirements for MIL-STD-23008B. Heats 1094, l103,and 1132 met the requirements specified in MIL STD 23008C. Heats 1103, 1118 and 1132 met the toughness requirements specified in MIL STD 23008D. Heats 1094 and 1118 had toughness values slightly below the minimum values specified in Revs. D and C respectively. In all cases, the measured values tended to be close to the minimum values specified in Revs. C and D.

While the experimental alloys did not exhibit toughness values greatly in excess of the minimum values specified fcr HY-80 in MIL-STD-23008 Revs. C and D, this should not be cause for concern at the present time. Since only a few heats have yet been produced, it is a certainty that there is a great deal more to learn regarding compositions and heat treatments and the effects of these on mechanical properties. It should also be pointed out that in heavy sections, cast HY-80, itself, does not appear to be capable of meeting the minimum toughness requirements of MIL-STD-23008D. In addition, if the experimental alloys are less prone to HAZ cracking than HY-80, as appears to be the case, perhaps slightly lower toughness levels would be acceptable.

Examination of the tempering curves in Figures 6 through 10 indicates that the experimental alloys begin reversion to austenite when heated to temperatures above about 1175° F. Thus, the first temper of the double-tempering treatment applied to these materials was actually an intercritical heat treatment. At the first (higher temperature) temper, a portion of the microstructure would transform to austenite while the remaining structure would undergo marked softening (and probably a large increase in toughness). Upon quenching from the first temper, the austenite which was present would then transform. Since cooling times from these relatively low temperatures would be short, The austenite would almost surely transform to martensite. The second temper would then serve to temper the newly-transformed martensite, thereby lowering the hardness and increasing the toughness of this material. By using the double tempering treatment, it was possible to attain lower hardness values then could have been achieved with single tempers. It is also likely that the double tempering treatment played a large role in producing the good toughness levels found in the experimental alloys.

The metallographic examinations seemed to indicate that in heavy sections, the experimental alloys attained microstructure consisting of martensite andlor lower bainite. It must be appreciated, however, that in low carbon steels, it is very difficult to distinguish optically between bainite andmartensite tempered at high temperatures. In order to fully characterize the microstructure of these alloys, it would probably be necessary to employ transmission electron microscopy. By performing dark-field imaging of the carbides, it should be possible to determine the exact nature of the microstructure.

The CCT diagram developed from the Gleeble evaluations appears to indicate a martensite start (MJ temperature of approximately 800° F. While this seems high for an MS temperature, the value could very well be this high because of the very low carbon level in the material. It appears that at cooling rates greater than about 6° F/second, the transformation product will be martensite. Lower cooling rates appear to result in transformation to bainite. Again, additional work would need to be performed in order to establish the nature of the transformation products in these alloys.

Regardless of the exact identity of the transformation products, the CCT diagram contains the features desired for the experimental alloy system. Ferrite and pearlite formation has been repressed to the extent that these transformation products do not occur even at a cooling rate corresponding to that of the center of a 30 in. thick cylinder in still air! For all practical purposes, formation of these toughness-impairing structures does not appear possible. Also, it appears that the bainite region hae been compressed to temperatures close to the MS temperature. This would promote a fine distribution of carbides which tends to give good toughness even if bainite does form. Given the CCT behavior indicated, it may be possible to reduce the nickel content of the experimental alloys slightly without detrimental effects. If so, this would be desirable since it would. make them less expensive.

The CCT diagram suggests a possible explanation for the higher impact toughness values exhibited by the 6 in. thick test blocks cast in the air-melt induction heats. It appears that these smaller test blocks would have transformed to martensite while the 12 in. thick test blocks of the AOD heats would have been at least partially bainftic. In general, martensitic microstructure will give better toughness.

The CCT diagram also gives an indication of the mechanical properties which could be expected in even larger sections than those evaluated in this investigation. It appears that for the maximum section thicknesses for which HY-80 is currently employed, the experimental alloys would probably give similar toughness levels to those found in the 12 in. thick blocks evaluated in this investigation.

Regarding weldability, the CCT diagram suggests that the HAZ'S of welds would probably display relatively good toughness over a wide range of cooling rates. With rapid cooling rates such as those which would occur with no preheating, the high temperature HAZ would probably 'transform to martensite. However, the apparent M_s temperature is quite high, allowing for substantial autotempering giving good toughness.

On the other hand, relatively low cooling rates such as those occurring with high interpass temperatures and high heat inputs would probably give much the same HAZ microstructure and properties. The present investigation, of course did not deal with this matter. Eowever, if funding for additional work is considered, a thorough study of the effects of cooling rate on HAZ properties would be valuable. If the experimental alloys are indeed tolerant of high heat inputs and interpass temperatures, this could possibly permit great reductions in welding costs.

The weldability tests revealed that the experimental alloys were quite resistant to hydrogen-assisted cracking of the HAZfs when welded without preheating using the processes and filler materials most commonly used to weld HY-80. None of the heats exhibited cracking in the CTS test for either the GMAW welds or the FCAW welds. In the much more severe Tekken tests, only two of the heats exhibited cracking in the FCAW tests. Collectively, this performance was much better than that given by the EY-80 specimens, which exhibited cracking regardless of welding process.

Since the Tekken test specimens are much more severely restrained than those used in the CTS test, the Tekken test was a better indicator of cracking susceptibility for these alloys. Also, the FCAW process always resulted in greater cracking susceptibility than did the welds deposited by GMAW. Because GMAW is a "hydrogen-free" process (if good workmanship is maintained), less cracking susceptibility than that developed with FCAW would be expected. As a result, the lower carbon heats 1103 and 1132 aPPear to be weldable without preheating despite the extreme severity of the Tekken test.

In the Tekken tests of the experimental alloys which did show cracking, the cracks initiated at the root of the weld and propagated through the weld metal (Fig. 26). Determining the exact initiation sites of these cracks is a difficult matter. It is possible that crack initiation did occur in the HAZ. However, it is at least as likely that the cracks initiated and grew solely in the weld deposit, which is, itself, susceptible to hydrogen-assisted cracking.

The cracking which occurred in the HY-80 welds had a much different appearance (Figure 27). In all cases, the HY-80 specimens exhibited classical "underbead" cracking of the HAZ. This would seem to indicate that cast HY-80 is far more susceptible to HAZ hydrogen cracking than the experimental alloys.

The measured diffusible hydrogen contents were fairly typical of the values usually found for the evaluated types of wire. This indicates that the apparent resistance of the experimental alloys to HAZ cracking was not due to abnormally low hydrogen levels in the weld deposits. The good results were evidently due to the nature of the alloys.

Comparing the microhardness surveys for welds in the experimental alloys with those for welds in HY-80 reveals the apparent cause for the better cracking resistance of the former materials. The maximum hardness in the HAZ's of the experimental alloys ranged from 320 to 360 HHN. The maximum hardness of the HAZ'S of the HY-80 welds were close to 500 KHN. It is well known that the hydrogen-assisted cracking susceptibility of steels increases markedly with increasing hardness. Clearly, when welded without preheating, HY-80 should be more susceptible to cracking.

It should be noted that throughout this investigation, whenever the term "no preheat" was used, it was intended to indicate that welding was done at room temperature (72°F). The experimental alloys would not necessarily exhibit good weldability at much lower temperatures. If temperatures were low enough to cause water condensation or even formation of ice on the material, it would not be expected that welding could be done without problems.

One of the intentions of Phase 2 of this investigation was to identify useful composition ranges and heat treatments for the experimental steels. While the scope of the present investigation did not permit a detailed study of these specific subjects, a few recommendations can be drawn from the results. It appears that appropriate ranges for composition would be as follows:

Table 17.

Recommended Composition Ranges for Experimental Alloy.
(Wt. Percent)

ELEMENT	RANGE
C	0.03 to 0.06
Mn	0.60 to 1.00
Si	0.25 to 0.50
Cr	1.25 to 1.75
Ni	5.2.5 to 5.75
Mo	0.40 to 0.60
S	0.010 max.
P	0.15 max.
Al	0.06 max.

Appropriate ranges for heat treating temperatures would probably be as follows:

Table 18.

Heat Treatment of Experimental Alloy.

~	TEMPERATURE (°F)	COOLING METHOD
NORMALIZE	1750 TO 1850	AIR COOL
HARDEN	1650 TO 1750	WATER QUENCH
FIRST TEMPER	1200 TO 1300	WATER QUENCH
SECOND TENPER	1075 TO 1150	WATER QUENCH

Should it be decided to produce the experimental alloys on a commercial basis, it. would almost certainly be necessary to employ AOD or VOD technology or ladle metallurgy techniques. These methods would be required to attain the low carbon levels in a cost-effective manner. These processes would also give the low hydrogen levels necessary to avoid hydrogen flakes in heavy sections. In addition, of course, they would also provide the low levels of sulfur, oxygent nitrogen and which promote good mechanical properties.

While the experimental alloys appear to be able to meet the mechanical property requirements of cast HY-80 and appear to be resistant to cracking when welded without preheating, it should be remembered that they are more highly-alloyed than HY-80. This will tend to increase steelmaking costs slightly. It is expected, however, that these increased costs would be more than offset by the reduced preheating requirements. In addition, the greater resistance to cracking, itself, should affect manufacturing costs favorably.

CONCLUSIONS

Test blocks from five air-melt induction heats (1089, 1090, 1091, 1092 and 1093) and four 2,000-lb AOD heats (heate #1094, #1103, #1118 and #1132) were heat treated and evaluated as potential replacements for HY-80. Based on initial mechanical properties and weldability tests using GMAW and FCAW, the following can be concluded:

1. The experimental alloys appear to be capable of meeting the mechanical property requirements for cast HY-80 per MIL-STD-23008B. The alloys show promise for being able to also meet the more stringent requirements of Revisions C and D of the specification. The alloys should be capable of attaining the required properties even when cast in heavier sections than those examined in the present investigation (12 in.).

2. The experimental alloys exhibit excellent resistanceto HAZ cracking when welded without preheating using the GMAW and FCAW processes and the filler materials used to weld HY-80. Cast HY-80 suffered severe HAZ cracking when welded under similar conditions.

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